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Association Notes

Causes of Hot Cracking

In this issue of the journal we are presenting three papers on this important subject:—

1. Tensile properties of mild steel weld metal at high temperatures, by J. G. Ball, B.Sc., A.I.M., and K. Winterton, Ph.D., B.Sc. (see page 104r).
2. Ductility of mild steel welds near the solidus, by E. C. Rollason, Ph.D., M.Sc., F.I.M. (see page 125r).
3. Explanation of hot cracking of mild steel welds, by E. C. Rollason, Ph.D., M.Sc., F.I.M., and D. F. T. Roberts, B.Sc.Tech. (see page 129r).

We are inviting a written discussion on these three papers, and should be glad to get early contributions, as we are hoping to publish the complete discussion in the February issue of the journal.

The first paper, by Ball and Winterton, describes an investigation in which tensile specimens cut from weld metal were heated in a platinum wound furnace, and tested to failure in a tensile testing machine in the range 900 deg. C. to 1100 deg. C. The mild steel weld metal specimens were found to have low ductility compared with the values at higher and lower temperatures. This brittle range should not be correlated with hot cracking, but it appears to be an effect general to cast steel. At higher temperatures (1200 to 1425 deg. C.) the weld metal was found to have very high ductility. The effects of segregation, direction of testing, prior heat treatment, welding current, rate of straining, iron oxide and sulphur content have all been investigated for their effects on the tensile properties.

The second paper, by Rollason, covers the same ground, though a much simpler testing procedure was used. A drop in ductility in the range 1250 to 1350 deg. C. was found in all cases, and this was thought to be connected with hot cracking. The results are correlated with those obtained from the Murex hot cracking machine. The author thinks that the difference in results compared with those obtained by Ball and Winterton is due to the fact that much higher rates of heating were used. Tests are reported for several electrode types and in particular for the effect of manganese additions in the electrode coating.

The paper by Rollason and Roberts gives an up-to-date account of the latest ideas on the causes of hot cracking. Hot cracking is stated to be largely due to the presence of sulphur, with manganese having an ameliorating increase. The silicon and phosphorus affect hot cracking in that they increase the segregation of sulphur. The authors also discuss the effect of pre-heating, welding current, and carbon and nickel content.

R.63

Tensile Properties of Mild Steel Weld Metal at High Temperatures

By J. G. BALL, B.Sc., A.I.M. and K. WINTERTON, Ph.D., B.Sc.

SYNOPSIS

Tensile specimens cut from weld metal were heated in a platinum wound furnace arranged to fit between the jaws of a tensile testing machine. No lagging was used in the furnace, reliance being placed in the vacuum and on reflecting inner surfaces to reduce convection and radiation losses respectively. The tests were made *in vacuo* to prevent oxidation and decarburisation.

In the range 900-1100 deg. C. all the mild steel weld metals tested were found to have low ductility compared with the values at higher and lower temperatures. As the loss of ductility was more pronounced in specimens cut transverse to the welding direction, and also as the ductility could be improved by annealing, it is probable that the effect was due to segregation. On reducing the pick up of impurities from the plate metal by buttering, the effect was diminished. Iron oxide in the deposit does not reduce ductility, and the effect of sulphur is uncertain. The present work has not shown that this region of low ductility is correlated with hot-cracking of weld metal, but it appears that it is an effect general to cast steel. It has been observed in a less marked degree in electrode core wire but not in wrought mild steel plate.

Mild steel weld metal tested at temperatures of 1200-1425 deg. C. has been found to have very high ductility. It has not been found practicable with the present apparatus to test mild steel at temperatures above the solidus, but a few tests on high carbon steel above the solidus showed that the tensile strength and ductility are not reduced to the very low values which might be expected.

Accounts have been appended of two complementary investigations: firstly, on the determination of the percentage and constitution of the oxide inclusions in three of the weld metals; secondly, on the examination of the structures of the same three weld metals under the electron microscope.

1. INTRODUCTION

This work originated as part of an investigation of the problem of hot-cracking in mild steel weld metal. The term hot-cracking refers to cracking occurring in weld metal at elevated temperatures, the surfaces of the crack being characteristically blued by oxidation. The consequences of hot-cracking are obvious but particularly important is its undermining of fatigue and impact strength. In addition it seems difficult to develop electrodes giving high strength deposits without introducing hot-cracking.

It seems likely that hot-cracking is due to some constitutional weakness in the weld metal at elevated temperatures. For instance, most workers have agreed that sulphur has a pernicious influence, and Reeve[1] has shown that hot-cracking can be induced by increasing the sulphur content of the electrode coating, the core wire or the plate metal. He has suggested that for a given set of welding conditions, there is a limiting sulphur content for the weld metal (sulphur limit), which when exceeded leads to hot-cracking. The role that sulphur plays in promoting hot-cracking is not understood, and no one has suggested that sulphur content is the only or even the prime factor involved. There is less agreement as to the influence of other factors—such as carbon, manganese, oxides and other inclusions, gases and state of deoxidation—though Reeve[1] has put forward some valuable evidence and suggestions.

Although welding may be regarded as a special type of casting operation, the unusual conditions make measurements difficult and most of the phenomena involved have not received much attention. For instance, there is little information on the transference of metal across the arc, arc temperatures, or the equilibria between metal slag and arc atmosphere. The rapid cooling of the weld metal (weld quench) prevents equilibrium being established in some of the complicated reactions which are probably similar to those known in steel making. Steep thermal gradients are set up in all directions in the weld, together with a complex system of stresses. The small amount of information does not permit the influence of these factors on hot-cracking to be assessed, particularly as the temperature at which the cracks occur has not been satisfactorily established, except in one special case involving very short runs where Antonioli[2] reported a temperature range of 1350-1000 deg. C. in which the cracks are formed.

It was decided to investigate the relation between mechanical properties and hot-cracking by determining the tensile properties of some mild steel weld metals, by reheating to various temperatures. The tensile specimens were cut from the weld deposits, and particular attention has been given to ductility.

There is not a great deal of published information on the tensile properties of steel at temperatures over 600 deg. C., and where this is available,[3,4,5,6,7] frequently most attention is given to tensile strength.[3,4] However, an exception must be made in the cases of Hall,[5] Dupuy[6] and Piwowarski,[7] who separately investigated the ductility of cast steel at high temperatures. Their results are discussed later.

Mention must be made of the work of Hodge[8] who describes results of tensile tests on weld metal. He is principally concerned with the effects of nitrogen at temperatures in the range of 20-600 deg. C., but he has done some work at higher temperatures.

2. EQUIPMENT

(a) Furnace Construction

The complete furnace unit is illustrated diagrammatically in Fig. 1 and pictorially in Fig. 2. In essentials it consists of Pt./Pt.-10 per cent. Rh winding supported on a mullite refractory tube (A, Fig. 1). The tube is wound for 2½ in. of its 3½ in. total length, and is of small internal diameter

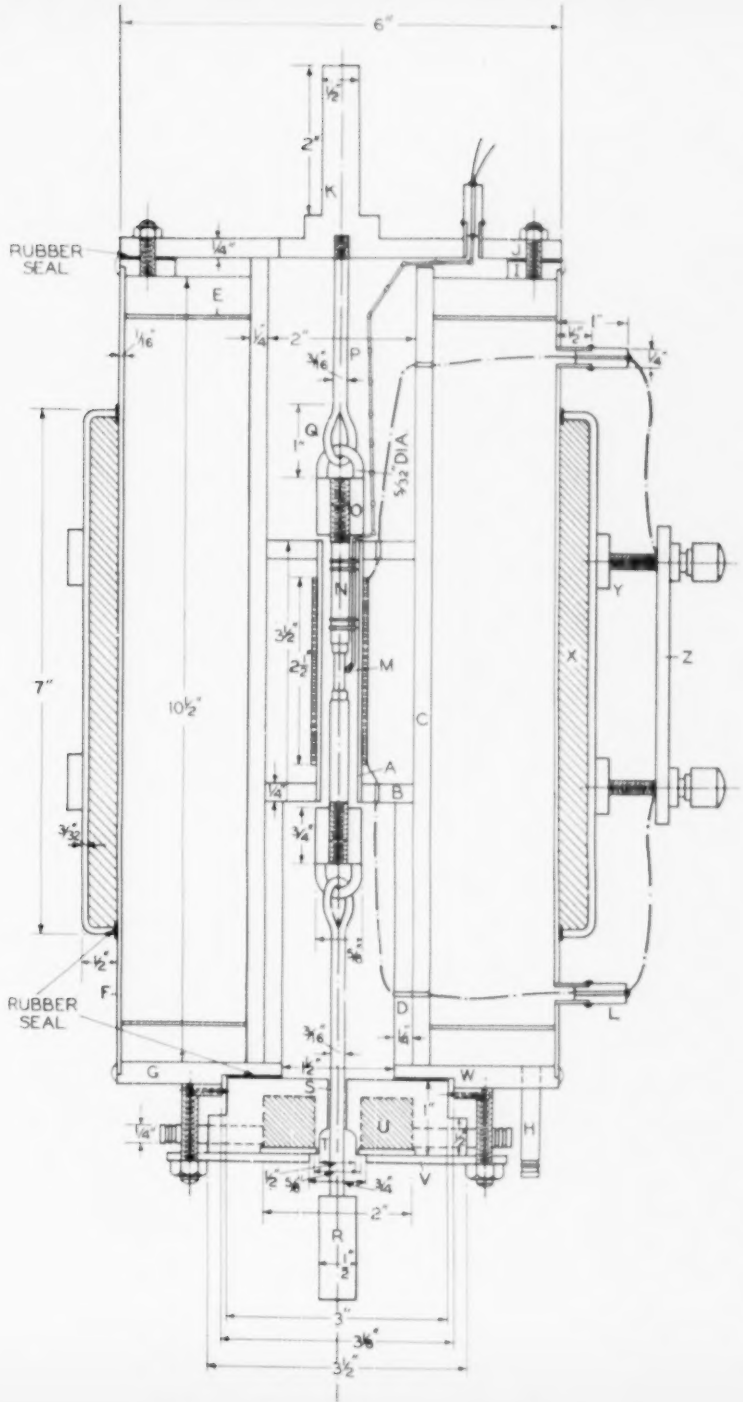

Fig. 1. High temperature furnace (for mechanical tests).

Fig. 2. The high temperature furnace unit

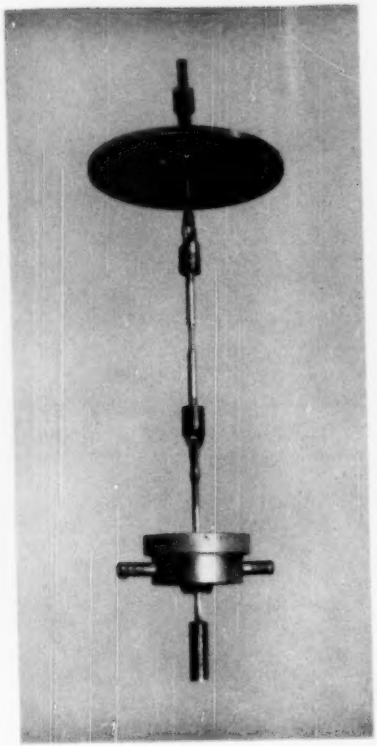

Fig. 3. The tensile mechanism.

to fit fairly closely round the specimen. This tube is fitted with mullite end plates (B) so that it rests inside a second refractory mullite tube (C) which is held in place by means of two discs (E) cut from mild steel. The furnace casing (F) and the supporting discs (E) are made from polished stainless steel to diminish radiation losses. At the lower end of the furnace a steel disc (G) is soldered to the casing. It is recessed to accommodate a brass gland, and carries a nozzle (H) which is normally connected to the evacuating system. At the upper end a steel flange (I) is soldered to the casing, and is provided with six bolts for fixing the top disc (J) in position; a rubber seal between these surfaces prevents loss of vacuum. The top disc has a centrally fixed stud (K) which is gripped in the upper jaws of the tensile machine. The windings leave the casing through small silvered porcelain bushes (L).

A Pt. Pt.–10 per cent. Rh thermocouple (M) is strapped to the specimen (N) so that the bare hot junction lies very close to the middle of the parallel portion without being in contact.

The tensile mechanism, shown separately in Fig. 3, consists of two plugs (O in Fig. 1) of Nimonic 80 screwed to take the ends of the specimen (N). These plugs are connected to thin rods of S.80 stainless steel by means of two stainless steel links (Q). The upper rod is screwed to the top plate (J) while the lower rod passes through a brass gland and is screwed into a plug (R) which is held in the lower grips of the tensile machine.

The brass gland shown in Fig. 1 was designed to have as little friction as possible while still maintaining a high vacuum when the parts are in relative movement during the tensile test. In essentials it consists of a reamed hole (S) fitting closely (with a clearance of 0·0005 in.) to the lower rod (P), and terminating in a grease cup (T) which is kept filled with a vacuum grease (Apiezon Grease M). To maintain the viscosity of this grease during the test the gland is water cooled, the water circulating round the space (U) as shown. The gland is held tightly against the bottom disc (G) by means of a clamp (V) which tightens a rubber seal between the machined surfaces of the disc and gland. Adjustment of the gland, to ensure axiality of the loading mechanism, can be made by means of the set screws (W); once set, the gland takes up the same position on subsequent occasions.

The furnace casing is a cooled by direct contact with water which circulates through a series of four brass compartments (X), held by steel straps (Y) to the casing against rubber seals.

A small terminal board (Z) is fixed to the side of the furnace and carries two pairs of terminals for the furnace windings and the thermocouple. The platinum winding is fed from a 230-80 volt 1·5 K.V.A. fixed ratio transformer supplied by a Variac transformer (continuously variable from 0-230 volts; 9 amp.; 1 K.V.A.). The purpose of this is to provide starting conditions of 37 amp. at 40 volts, falling to 12·5 amp. at 40 volts at 1500 deg. C.

(b) Furnace Design

The unusual procedure has been adopted of having no furnace lagging, since, as the furnace is to work *in vacuo*, convection is at a minimum; radiation losses have also been reduced by introducing reflecting inner surfaces of polished stainless steel as far as possible. The most serious loss of heat envisaged was by conduction away from the specimens along the metallic grips to the tensile machine, and this was minimized by introducing links of austenitic steel in series to give a small area of contact in low conductivity material.

Precautions were taken in the design and during the construction of the furnace to prevent non-axial loading. The outer surfaces of the flange, (I) in Fig. 1, and the recess of the bottom disc (G) which fit against the rubber seals of the top disc and the brass gland respectively, were machined as a single unit after fixing to the furnace casing in order to ensure that these surfaces were parallel. The brass gland was fixed in a central position opposite the top stud by means of the adjusting screws (W); once these are set no further adjustment is required. These precautions, together with the point contact provided by the austenitic links, ensure good alignment.

The brass gland described was designed after a little experimentation, and has been found satisfactory in having low frictional resistance (in practice the frictional load was not measurable on the testing machine) and in maintaining a high vacuum while parts are moving during the test.

(c) The Evacuating System

A Speedivac—type 2 high vacuum pump was used. This is a two-stage mechanical pump giving a nominal vacuum of 0·00001 mm. mercury and a displacement of 48 litres per minute. The pump is connected to the furnace through a drying tube containing phosphorus pentoxide. A McLeod gauge (range 2 mm.-0·001 mm. mercury) is connected in parallel with the furnace for the measurement of low pressure.

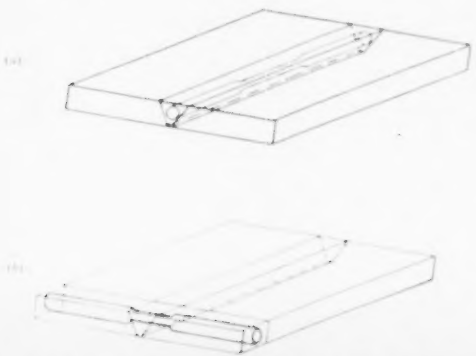

Fig. 4. Position of tensile specimens cut from welded joints.
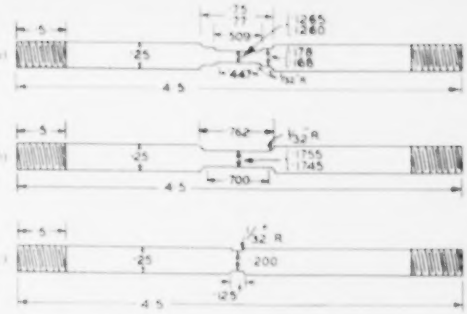
Fig. 5. Dimensions of specimens for tensile tests.

3. GENERAL METHOD

(a) Preparation of the Tensile Specimens

The all-weld metal specimen is cut from a butt-weld made in two mild steel plates as shown in Fig. 4 (a). The 60 deg. "V" is filled with weld metal deposited from the electrode under investigation using two or three runs made under standard conditions. A test specimen is then machined to the dimensions shown in Fig. 5(a) or 5(b). In the early stages of the work the type shown in Fig. 5(a) was used, but when it was found that the screw threads were strong enough to permit a greater diameter in the parallel portion, the shape shown in Fig. 5(b) was preferred.

Some specimens have been cut at right angles to the welding direction as shown in Fig. 4(b), to ascertain the directional nature of the mechanical properties. To avoid fracture in the plate metal, specially shaped specimens had to be cut (Fig. 5(c)) in which the parallel portion was very short and within the weld.

(b) Test Procedure

The test specimen, after measurement of diameter and gauge length by means of a travelling microscope, is screwed to the upper Nimonic plug, (O) in Fig. 1. The thermocouple is strapped in position, and the top end of the furnace is bolted and sealed. The other Nimonic plug and the lower rod are then screwed to the specimen, and the brass gland is slipped over the rod and clamped in position. Finally the lower plug (R) in Fig. 1 is added.

Evacuation is then commenced, and the water-flow through the baths is adjusted to a suitable value. When the vacuum reaches about 0·003 mm. mercury, the temperature is raised rapidly to the required level. The time necessary for heating is about 10-15 minutes to reach temperatures of 1000-1100 deg. C. A period of five minutes at temperature is allowed for the specimen to attain uniformity, after which it is tested to fracture.

The pressure found to be satisfactory at the commencement of the test was 0·003-0·007 mm. mercury; with the furnace at high temperatures the pressure that could be maintained was slightly higher (0·01-0·015 mm. mercury), due mainly to increased gassing of refractories. Scaling and decarburisation were very slight under these conditions.

After the test, cooling takes place in continued *vacuo* and the apparatus can be dismantled when non-scaling temperatures are reached. Cooling is rather slow, owing to the design of the furnace, which tends to conserve heat, so

a few attempts have been made to cool the contents in a stream of pure dry nitrogen. This method was abandoned because it led to over-heating of the contents by convection heating; slight scaling resulted either because the convection heating caused increased gassing of the refractories, or because the nitrogen was insufficiently pure. The method of cooling *in vacuo* has therefore been preferred.

When cool, the specimen can be removed, and its reduction of area and percentage elongation can be determined by a travelling microscope, followed where desirable by microscopical examination.

(c) Special Conditions of Testing

After the period of five minutes soaking at temperature, the temperature is fairly uniform over the gauge length. Measurements showed a variation of temperature between the centre of the gauge length and the shoulder amounting to 2 deg. C. at 800 deg. C. and 5 deg. C. at 1100 deg. C. At higher temperatures the variation may be rather more, and for this reason reduction in area is the more reliable indication of ductility.

At high temperatures, above 1,000 deg. C., a difficulty arose in that welding took place in the screw threads between the specimen and the Nimonic plugs. This was overcome by lightly dusting the screw threads with alumina powder.

The Dennison testing machine used in these tests may be operated at four constant speeds; in practice the highest speed was used corresponding to 0·8 in. per min. The straining rate was therefore approximately constant and equivalent to about 0·025-0·015 per sec., depending on the extension.

4. MATERIALS

Two plate steels have been used (steels 1 and 2), with analyses as shown in Table I, the most important difference between them being in sulphur content.

Tests have been made on a high carbon steel (steel 3), and for convenience its analysis is included in Table I.

The electrodes used are listed with their B.E.A.M.A./I.O.W. classification numbers in Table I, together with the analysis of their weld deposits for the two types of steel plate. The properties of the weld metals from two of these

Table I. Analysis of Steels Used

Mark	Class	C	Si	Mn	S	P
1	—	0·18 0·19	0·038	0·52	0·070 0·073	0·030
2	—	0·13	Trace	0·56	0·041	0·035
3	—	0·85	0·20	0·64	0·028	0·034
Analysis of Weld Deposits on Steel 1						
A	E 217	0·14 0·14	0·127	0·39	0·060 0·062	0·024
B	E 217	0·10 0·10	0·122	0·062	0·044	0·040
C	R 416	0·10 0·10	0·041 0·042	0·35	0·062	0·030
D	E 537 *	0·07 0·07	0·028	0·09 0·10	0·034	0·024
Analysis of Weld Deposits on Steel 2						
A	E 217	0·08 0·09	0·085	0·37 0·38	0·045	0·025
B	E 217	0·07 0·07	0·16	0·64 0·64	0·031	0·043
C	R 416	0·10 0·10	0·062	0·33	0·037	0·028
D	E 537 *	0·06 0·06	0·024	0·09 0·09	0·029 0·030	0·011 0·010

* This electrode is of the type E 537 of the B.E.A.M.A./I.O.W. classification, though it was not included at the time these experiments were made.

electrodes have been extensively compared, electrode A being chosen as being susceptible to hot-cracking, and electrode B as being relatively free from the defect.

Electrodes C and D having an intermediate and high oxygen content respectively were chosen to investigate the influence of oxygen since electrodes A and B give deposits of low oxygen content.

The remaining electrode (electrode E) is an austenitic steel rod of the 18/8 type.

5. RESULTS

In considering the properties of weld metals at high temperatures in relation to hot-cracking, ductility has been considered of prime importance. Fundamentally the value of ductility decides whether the weld metal can sustain without fracture the self-imposed thermal contractions.

Reduction in area is considered a better index of ductility than percentage elongation, one reason being that it is less liable to error through thermal gradients in the tensile specimen. The view has been justified in practice as reduction in area has given more reliable and consistent results than percentage elongation.

It is not easy to group the investigations which follow into rigid sections, though it may be helpful to attempt this.

(i) Two weld metals were chosen, one free from, and the other prone to hot-cracking, and their ductility was compared over a wide range of temperature. A region of low ductility was found at temperatures of 900-1000 deg. C. being rather more marked for the electrode prone to hot-cracking (section (a)).

(ii) Attempts were made to correlate the region of low ductility with susceptibility to hot-cracking. (Sections (b), (c) and (d).)

(iii) Investigations were made to explain the region of low ductility in metallurgical terms. (Sections (e) to I inclusive.)

(iv) The region of low ductility was found to be a general phenomenon of cast steel. Some consideration has been given to the significance of the results at higher temperatures. (Sections (m), (n) and (o.)

(a) Ductility of Two Weld Metals

In the first instance, two electrodes were chosen for examination:—

Electrode A—susceptible to hot-cracking.
Electrode B—relatively free from the defect.

The preliminary results from deposits from these electrodes are given in Tables II and III. Fig. 6 shows the results for

Table II. Tensile Properties of Weld Metal (Electrode A/Steel 1) (Longitudinal specimens (Fig. 4a); dimensions as shown in Fig. 5a)

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Elongation per cent.	Reduction in Area per cent.
1	710	12·7	46	86
2	825	7·8	58	88
3	915	6·0	44	41
4	970	5·5	48	49
5	1020	4·45	52	60
6	1130	2·9	65	84
7	1223	2·3	73	98·5
8	1318	2·2	81	100
9	1350	1·5	72	100
10	1400	0·9	42	88
11	1425	—	67	100

Table III. Tensile Properties of Weld Metal (Electrode B/Steel 1) (Longitudinal specimens (Fig. 4a); dimensions as shown in Fig. 5a)

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Elongation per cent.	Reduction in Area per cent.
12	705	13·5	36	70
13	756	11·2	43	89
14	825	7·0	68	90
15	915	4·7	31	67
16	920	4·8	52	69
17	935	5·2	46	69
18	958	5·6	39	58
19	970	6·3	—	57
20	1020	4·2	61	68
21	1020	4·2	63	70
22	1100	3·2	49	84
23	1223	2·4	72	96
24	1318	1·6	89	100

reduction in area plotted as a function of temperature. For specimens cut entirely from weld metal (Fig. 4(a)), the curves for weld metal from electrode A and electrode B are shown in thin line and broken line respectively. The two curves show similar features, namely, a minimum in ductility at 900-1000 deg. C., and very high ductility at temperatures above 1200 deg. C.

The minimum in ductility is more marked with the electrode prone to hot-cracking, and appears at a lower temperature. Since the temperature at which hot-cracking occurs was not known, the relevance of these results to hot-cracking was uncertain, and it was decided to make a fuller investigation of the phenomenon.

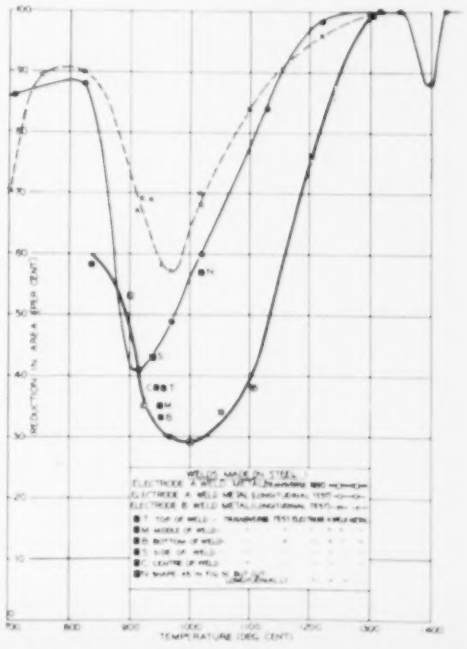

Fig. 6. Ductility of two weld metals; effect of direction of testing and position of specimen.

(b) Effect of Direction of Testing on Ductility

It seemed clear that the minimum in ductility referred to above was associated with some constitutional effect in the weld metal. Welds are not symmetrical from the point of view of major segregation; any impurities would tend to be segregated by columnar crystal growth to a central plane parallel with the welding direction.

Consequently it seemed worth while to test the mechanical properties at right angles to the welding direction. This involves the use of a composite weld-plate specimen, and to ensure that fracture occurred in the weld, the shape shown in Fig. 5(c) was found necessary. Here a very short parallel portion is used and the percentage elongation is not measured.

The results for electrode A weld metal are given in Table IV, and the curve for reduction in area is shown in Fig. 6 (thick line). This curve shows similar features to those obtained from all weld metal specimens, but there is a lower value for minimum ductility and this occurs at a higher temperature.

Specimens of this shape might introduce the brittleness associated with notching. A specimen was therefore cut to this shape (Fig. 5(c)) entirely from weld metal, i.e. in the direction of welding. The results are given in Table IV (Test 30), and the reduction in area is plotted as point N in Fig. 6. The joint lies very close to the curve for all weld metal specimens, cut to the shape shown in Fig. 5(a). Similar results were obtained in tests on wrought mild steel (points Y and Z in Fig. 14).

Evidently the changes due to specimen shape are small and may be disregarded. It follows therefore that the different shapes of the two curves for electrode A weld metal are associated with the difference in the direction of testing, i.e. to anisotropy of the weld metal.

A few results have been obtained on specimens cut transversely from electrode B weld metal, and are reported in Table IV. These are insufficient to give a reliable curve for ductility, but they show a similar tendency to that found with electrode A weld metal.

(c) Effect on Ductility of Position of Specimen within the Weld

The anisotropy of the weld metal suggested a closer investigation into the possibility of segregation of impurities within the deposit. Three tests were made on small specimens cut transversely, one above the other, from a single weld. Tests were made at a temperature of about 950 deg. C. and the results are given in Table V. The ductility values are plotted as points T, M and B in Fig. 6, where it may be seen that the ductility of the specimen from the bottom of the weld is less by some 5 per cent. reduction in area than that for the specimen from the top of the weld. While the total effect is small, its trend is unexpected, in that the influence of segregation would be supposed to be most marked at the top of the weld.

Further tests were made to determine whether there were segregation effects from the side to the centre of the weld. Two small transverse specimens were cut from adjacent sections in a wide desposit, and machined to give a short gauge length lying within the weld; in one specimen the gauge length was at the side of the weld, and in the other at the centre. The results are given in Table V, and the ductility values are plotted as points S and C in Fig. 6. The difference in ductility between the two specimens is again small, but it suggests that there is some segregation of impurities towards the centre of the weld.

Table IV. Tensile Properties of Two Weld Metals

(Transverse specimens (Fig. 4b); dimensions as in Fig. 5c)

Weld Metal (Electrode A/Steel 1)

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq in.)	Reduction in Area per cent.
25	836	9·6	58
26	900	6·0	53
27	923	6·5	35
28	967	5·6	30
29	1000	5·1	29
30*	(1024)	(4·7)	(57)
31	1051	4·6	34
32	1108	3·9	38
33	1205	3·0	76
34	1307	1·8	99·5

Weld Metal (Electrode B/Steel 1)

35	825	6·9	77
36	961	6·7	41
37	993	5·9	48
38	1065	4·4	45
39	1303	1·9	98

* This specimen was shaped as shown in Fig. 5c, but cut longitudinally. It is plotted as point N in Fig. 6.

Table V. Tensile Properties of Weld Metal at Various Points within the Weld (Electrode A/Steel 1)

(Transverse specimens (Fig. 4b); gauge dimensions 0·1 in. dia. × 0·125 in. length)

Test No.	Position*	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq in.)	Reduction in Area (per cent.)
40	T (Top)	953	6·0	38
41	M (Middle)	950	6·6	35
42	B (Bottom)	952	7·3	33
43	S (Side)	939	5·6	43
44	C (Centre)	945	5·6	38

* The letters in this column correspond with those shown in Fig. 6.

In these tests (tests 40-44, Table V) it will be noted that none of the ductility values lies on the curve for the normal transverse specimens; this may be due to the fact that smaller specimens were used for these special tests.

(d) Effect of Welding Current on the Tensile Properties of Weld Metal

The weld metal used for mechanical tests was normally deposited at 170 amp. for 8-gauge mild steel electrodes. Since welding current greatly influences the hot-cracking tendency, it was felt that some similar effect might be reflected in the tensile properties.

Tests were therefore made on weld metal from electrode A deposited at 230 amp., and the results are reported in

Table VI. The ductility values are plotted in Fig. 7. The curve is displaced and suggests that the higher welding current has an annealing effect. The higher welding current appears to have less effect on longitudinal specimens, though only a few tests were made (see Table VI).

Table VI. Tensile Properties of Weld Metal (Electrode A Steel 1) Deposited at a High Welding Current (230 amp.)

(Transverse specimens (Fig. 4b); dimensions as shown in Fig. 5c)

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Elongation per cent.	Reduction in Area per cent.
57	838	6·8		68
58	899	6·6		39
59	932	6·5		34
60	999	5·1		49
61	1108	3·3		79
62	1215	2·1		96

(Longitudinal specimens (Fig. 4a); dimensions as shown in Fig. 5a)

63	851	5·1	60	83
64	897	5·2	36	37
65	960	5·8	34	48

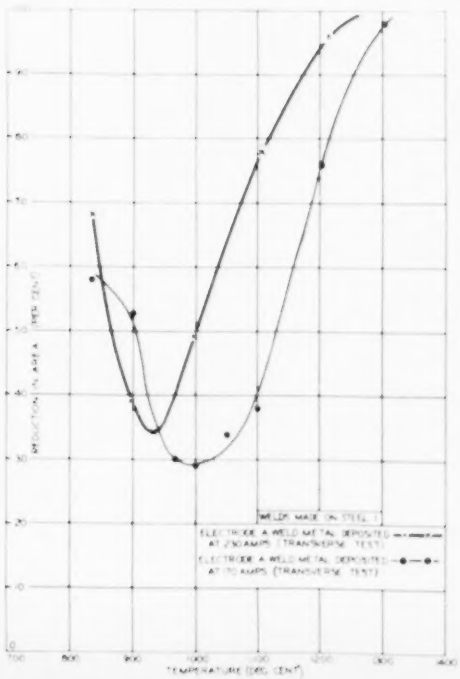
REDUCTION IN AREA PER CENT.
TEMPERATURE (DEG. CENT.)
WELDS MADE IN STEEL 1
ELECTRODE A WELD METAL DEPOSITED AT 230 AMPS (TRANSVERSE TEST)
ELECTRODE A WELD METAL DEPOSITED AT 170 AMPS (TRANSVERSE TEST)

Fig. 7. Effect of welding current on ductility.

(e) Effect on Ductility of Heat-Treatment before Testing

As previously explained, when heating the specimen before testing, a period of five minutes is allowed at the testing temperature to ensure uniformity of temperature over the gauge length. Two tests were made in which this five-minute soaking period was omitted, and the results are reported in Table VII (tests 45 and 50). The ductility values are plotted as points F and L in Fig. 8. Evidently the normal soaking period is desirable, the effect of its omission being as expected.

In explaining the general shape of the curves so far obtained, it was thought possible that:

(i) The minimum in ductility is due to minor segregation of some impurity on the grain boundaries.
(ii) This is accentuated due to major segregation in transverse specimens.
(iii) The effect is increasingly diminished with rising testing temperature by diffusion and homogenisation.

To test the third suggestion, several tests have been made in which various heat treatments were applied before testing. The results are reported in Table VII and the ductility values are plotted in Fig. 8.

Four tests were made in which the normal soaking period of five minutes was extended to twenty minutes. In all cases the ductility was improved, though as might be expected the change was smaller in the heavier section transverse specimens (points M and N in Fig. 8) than in the longitudinal specimens (points G and H).

Next a test was made in which a period of four minutes, soaking at 1200 deg. C. was applied, the temperature then being reduced rapidly to about 925 deg. C., and the test made at the lower temperature in the usual way (point J in Fig. 8). The prior treatment had an adverse effect on ductility. This was surprising, but a repeat test confirmed the result (point K in Fig. 8); the effect remains unexplained. In a similar test in which prior treatment was given for four minutes at 1340 deg. C. before testing at about 1000 deg. C., a considerable improvement in ductility was effected (point R in Fig. 8).

Three further tests were carried out to investigate the effect of soaking at temperature, but the specimens were annealed at 1260 deg. C. for four hours in a separate furnace. The specimens were allowed to cool to room temperature between the annealing treatment and the reheating for test purposes. The ductility values are plotted as points O, P and Q in Fig. 8. Evidently the annealing treatment has considerably improved the ductility.

With one exception, annealing before testing has generally the expected beneficial effect on ductility. However, this would only partly explain the elimination of brittleness at high testing temperatures.

(f) Dilatometric Investigation of Two Weld Metals

Consideration was given to the possibility that loss of ductility at approximately 900-970 deg. C. was connected with the Ac_3 change. For this purpose expansion temperature curves for weld metals deposited from electrodes A and B were determined under heating conditions using a Leitz dilatometer. The critical temperatures were approximately the same for both weld metals, the Ac_3 point occurring at 905 deg. C., and the Ac_2 point at 720 deg. C. The difference in hot-cracking between the two electrodes cannot be attributed to any difference in the temperature of the critical range, and further, this critical range does not coincide with the range of low ductility.

Table VII. Tensile Properties of Weld Metal Heat Treated before Testing (Electrode A/Steel 1)

Test No.*	Prior Heat Treatment	Position of Specimen	Shape of Specimen	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Elongation (per cent.)	Reduction in Area (per cent.)
45 (F)	Rapid heating, no soaking	As in Fig. 4a	As in Fig. 5a	894	5·3	46	48
46 (G)	20 mins. at 918 deg. C.	..	..	918	6·4	41	44
47 (H)	20 mins. at 937 deg. C.	..	..	937	6·1	50	58
48 (J)	4 mins. at 1200 deg. C.	..	..	923	5·4	27	34
49 (K)	4 mins. at 1200 deg. C.	..	..	928	5·4	35	34
50 (L)	Rapid heating, no soaking	As in Fig. 4b	As in Fig. 5c	1183	—	—	60
51 (M)	20 mins. at 923 deg. C.	..	..	923	6·7	—	36
52 (N)	20 mins. at 1109 deg. C.	..	..	1109	3·8	—	45
53 (O)	3 hrs. at 1260 deg. C.	..	..	919	7·3	—	50
54 (P)	3 hrs. at 1260 deg. C.	..	..	1011	5·8	—	46
55 (Q)	3 hrs. at 1260 deg. C.	..	..	1108	3·8	—	100
56 (R)	4 mins. at 1340 deg. C.	..	..	1006	4·7	—	81

* The bracketed letters in this column correspond with those in Fig. 7.

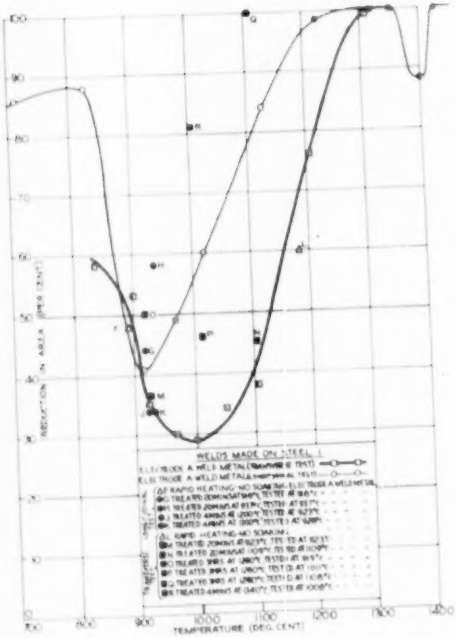

Fig. 8. Effect of prior heat treatment on ductility.

(g) Effect of Iron Oxide Content of the Weld Metal on its Ductility

Dr. L. Reeve drew the authors' attention to some work by A. Niedenthal[9], showing that on some instances, oxygen content has an undesirable influence on the strength and ductility of steel tested at high temperatures in the impact bend test.

Iron oxide content was considered to be a more significant variable than total oxygen content, the latter being merely a summation of the various oxide contents. It is not possible to devise a strictly valid method for investigating the effect of iron oxide content, since this factor cannot be isolated and inevitably involves other changes in the deposition and composition of the weld metal. To discover any outstanding effect of iron oxide content, three electrodes were chosen which were supposed to vary widely in the amount of iron oxide contained in the deposits. Electrodes C and D were chosen because in their deposited weld metal they were stated to give respectively intermediate and high iron oxide content compared with either electrode A or B.

Samples of three of the deposited weld metals were submitted to the National Physical Laboratory for a complete examination of the non-metallic residues. Their report on this examination is included as Appendix 2 of the present paper. In the table which follows, results are included for analyses for electrode B weld metal, these being taken from an earlier report from the same laboratory. It will be seen that the total oxygen contents are in the expected order, though the iron oxide content for electrode C is actually lower than that for electrode A. The method consisted in making an alcoholic iodine extraction followed by analysis of the residues by microchemical and X-ray diffraction methods. The following figures are selected from the results:

Deposit	FeO	MnO	SiO_2	Total Oxygen
Electrode A	0·013	0·084	0·088	0·082
Electrode B	0·0055	0·088	0·081	0·075
Electrode C	0·009	0·164	0·082	0·095
Electrode D	0·700	0·010	0·003	0·163

The chemical compounds present were determined to be as follows:

Electrode A—A high silica content SiO_2, though this was difficult to identify,* together with some manganese sulphide.

Electrode B—Rhodonite $(MnFe)O \cdot SiO_2$ + manganese sulphide MnS.

Electrode C—A complex silicate $2(Mn \cdot Fe)O \cdot SiO_2$. No sulphides were detected.

Electrode D—Iron oxide (FeO) together with some Fe_3O_4. Iron sulphide (FeS) was present.

It should be noted therefore that there can only be traces of free iron oxide present in the deposits from electrodes A, B and C, and that with electrodes B and C the iron oxide is present in combination with silica as complex silicates. It is only with electrode D that substantial amounts of free iron oxide are present in the deposit.

It is now possible to consider the significance of these analyses in relation to the mechanical test results. Mechanical tests were made on all weld metal specimens, cut longitudinally from electrode A, C and D, using steel 2 plate metal. The results are reported in Tables VIII, IX and X, and the ductility values are plotted in Fig. 9.

The value for minimum ductility and its corresponding temperature is different for each type of deposit. The lowest value for minimum ductility is associated with the highest temperature with the lowest iron oxide content. In the case of electrode D, known to contain a large amount of free iron oxide, the minimum in ductility is not quite so pronounced.

It appears that there is little relation between the iron oxide content and the values of ductility in the temperature range 900-1100 deg. C.

(h) Effect of Sulphur Content of the Plate Steel on the Properties of Weld Metal

For most of the investigations plate steel 1 was used, and this was found to have a rather high sulphur content (0·070-0·073 per cent.). As a result the weld metal deposited on this steel will also have a fairly high sulphur content (0·060-0·062 per cent. for electrode A). On the other hand, in order to avoid possible complications, the tests on iron oxide content were made using mild steel plate of relatively low sulphur content (0·041 per cent.).

To examine the effect of sulphur content on the properties of the resultant weld metal, we may compare the results obtained on weld metal from electrode A deposited on steels 1 and 2. In addition, further tests were made on weld metal from electrodes C and D on steel 1, for comparison with the results on steel 2. These results are reported in Tables II, VIII, IX and X, and the ductility values are plotted in Figs. 10, 11 and 12.

It will be observed that the investigation has been mainly restricted to the effect of sulphur content on the mechanical properties in the region of minimum ductility. It is not easy to summarise the results which are in fact conflicting, but on the whole the change in sulphur content does not seem to affect ductility greatly at these temperatures.

(i) The Effect of Using "Padded" Plates on the Tensile Properties of Weld Metal

Weld metal composition is partly determined by the composition of the plate metal. In practice cracking troubles have often been overcome by lining the surfaces of the plates to be welded with high-grade steel (weld metal) before

* Possibly due to combination of titania with the silica in the form of a complex titano-silicate (see Appendix 2).

Table VIII. Tensile Properties of Weld Metal (Electrode A Steel 2)

(Longitudinal specimens (Fig. 4a); dimensions as shown in Fig. 5b)

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Elongation per cent.	Reduction in Area per cent.
66	814	7·7	52	89
67	921	4·8	63	68
68	1007	5·3	36	41
69	1058	4·3	40	52
70	1106	3·6	49	61
71	1199	2·6	67	98

(Transverse specimens (Fig. 4b); dimensions as shown in Fig. 5c)

72	922	6·7	—	36
73	997	5·6	—	29
74	1003	4·6	—	36
75	1049	4·9	—	45
76	1105	4·2	—	47

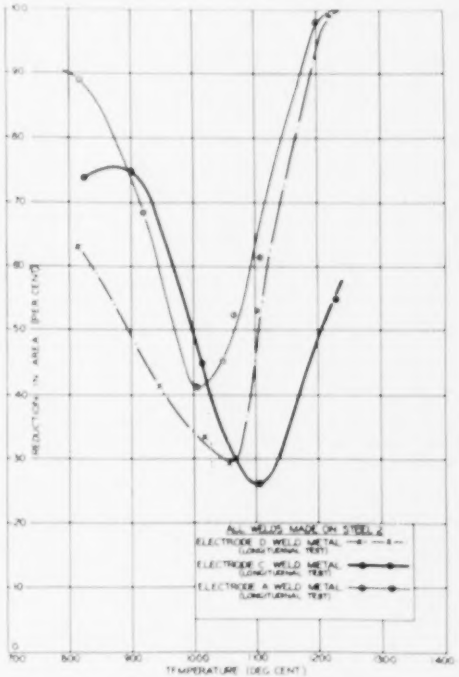
REDUCTION IN AREA (PER CENT.)
TEMPERATURE (DEG. CENT.)
ALL WELDS MADE ON STEEL 2
ELECTRODE D WELD METAL (LONGITUDINAL TEST)
ELECTRODE C WELD METAL (LONGITUDINAL TEST)
ELECTRODE A WELD METAL (LONGITUDINAL TEST)

Fig. 9. Ductility of three weld metals with various oxygen contents.

Table IX. Tensile Properties of Weld Metal from Electrode C Deposited on Two Different Plate Steels

(Steel 1 Longitudinal Specimens (Fig. 4a); dimensions as in Fig. 5b)

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Elongation per cent.	Reduction in Area per cent.
77	1004	5·9	43	56
78	1051	4·5	32	46
79	1108	4·5	31	39
80	1157	3·5	37	49

(Steel 1 Transverse specimens (Fig. 4b); dimensions as in Fig. 5c)

81	927	7·4	—	50
82	1003	6·3	—	42
83	1045	5·4	—	42
84	1111	3·9	—	40
85	1147	3·8	—	53

(Steel 2 Longitudinal specimens (Fig. 4a); dimensions as in Fig. 5b)

86	824	5·6	56	74
87	900	4·0	54	75
88	1013	5·8	39	45
89	1104	4·5	25	26
90	1228	2·7	45	55

Table X. Tensile Properties of Weld Metal from Electrode D Deposited on Two Different Plate Steels

(Steel 1 Longitudinal specimens (Fig. 4a); dimensions as shown in Fig. 5b)

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Elongation per cent.	Reduction in Area per cent.
91	819	4·7	33	49
92	919	2·6	27	37
93	1000	3·1	20	24
94	1064	3·8	23	28
95	1112	3·3	62	90

(Steel 2 Longitudinal specimens (Fig. 4a); dimensions as shown in Fig. 5b)

96	819	8·5	42	63
97	946	—	30	41
98	1018	4·3	24	33
99	1058	3·9	25	29
100	1106	3·4	32	53
101	1219	2·1	70	99

the final joint is made. By this means plate dilution is reduced from about 30 per cent. to some much lower figure.

Tests were made on specimens cut from electrode A weld metal deposited on plates "padded" with electrode B weld metal (lining about $\frac{1}{4}$ in. thickness). The results are reported in Table XI, and in Fig. 13 the ductility values are compared with those from electrode A weld metal on untreated plates.

The two curves are similar in shape, padding causing a displacement in the direction of greater ductility. The effect is not very marked, which suggests that the principal impurities responsible for lack of ductility are not derived from the plate metal. The tensile strength is also higher when padded plates are used, and this necessitated a slightly smaller gauge diameter than that shown in Fig. 5(c).

Evidently dilution of the weld metal with plate metal has no marked effect but causes a slight loss of ductility.

(j) Ductility of Two Core Wires and Wrought Mild Steel

Small specimens were cut for test from the core wires of 8-gauge electrodes A and B. The results are reported in Table XII and the ductility values are plotted in Fig. 14 where the corresponding values for the respective weld metals are included for comparison. The ductility for the core wire is in general much higher than that of the corresponding weld metals. The values for reduction in area are higher, for testing temperatures in the range 800-1100 deg. C., in the electrode A core wire, though the reverse is the case for the respective weld metals. It appears that low ductility in the weld metals in this temperature range is a result of melting during the welding process, though it should be remembered that the composition of the weld metal is partly controlled by the electrode coating.

The material of core wires is wrought mild steel, generally a rimming steel. For comparison tests have been made on some rolled steel—steel 1—the results being reported in Table XIII, and the ductility values included in Fig. 14. Ductility values are even higher with this steel in the same temperature range, despite a relatively high sulphur content (0·070-0·073 per cent.).

(k) Tensile Strength of Weld Metal and Wrought Mild Steel

In these investigations the main emphasis was on ductility though tensile strength was recorded. In Fig. 15 two curves are plotted for electrode A and electrode B weld metal; they show the same trend—a progressive decay of strength with increasing temperature. Some relatively high values were obtained in the critical range 900-1000 deg. C. for electrode B weld metal. Tensile strength values for wrought mild steel are included for comparison.

Around 900 deg. C. the values for tensile strength are somewhat higher than published values for cast mild steel. Dupuy[6], for example, found a strength of 4·3 tons per sq. in. at 900 deg. C. declining to 2·3 tons per sq. in. at 1100 deg. C. for a mild steel containing 0·15 per cent. carbon. This lower strength is explained by the smaller strain rate which he used (0·002 per second). At higher temperatures, Hall[5] found a maximum tensile strength of 1·5 tons per sq. in. at 1300 deg. C., falling to 0·9 tons per sq. in. at 1400 deg. C., for a mild steel containing 0·10 per cent. carbon. Here the straining rate is more comparable (0·053 per second, compared with 0·02 per second in the present investigation) and the figures are in good agreement with the strength values obtained for weld metal.

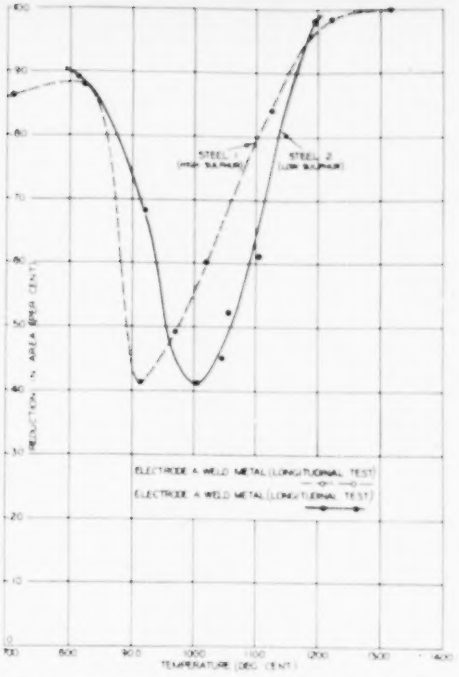

Fig. 10. Effect of sulphur content of plate steel on ductility.

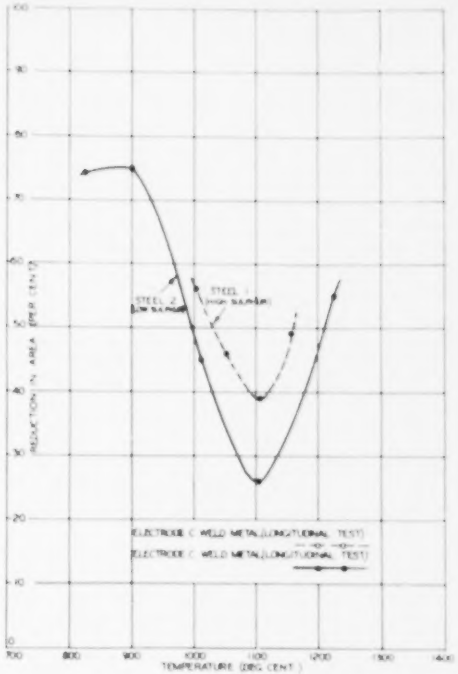

Fig. 11. Effect of sulphur content of plate metal on ductility.

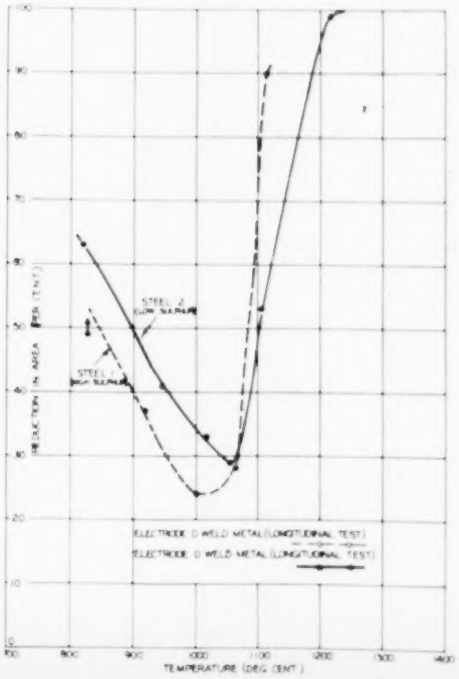

Fig. 12. Effect of sulphur content of plate metal on ductility.

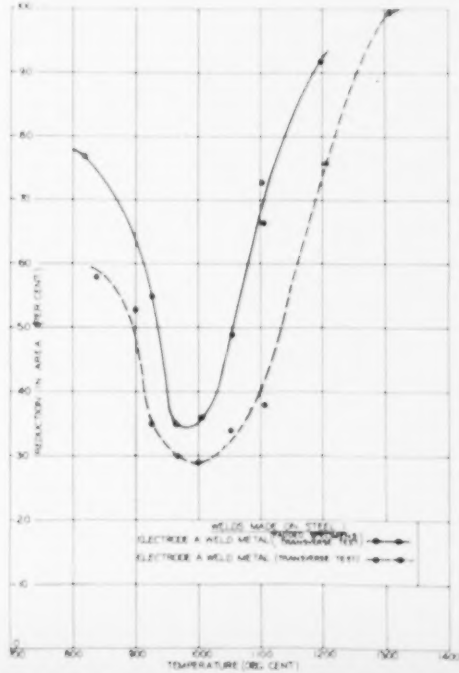
WELDS MADE IN STEEL
(HIGH SULPHUR)
ELECTRODE A WELD METAL (TRANSVERSE TEST)
ELECTRODE A WELD METAL (TRANSVERSE TEST)
REDUCTION IN AREA (PER CENT)
TEMPERATURE (DEG. CENT.)

Fig. 13. Effect of "padded" plates on the ductility of weld metal.

(l) Effect of Rate of Straining on the Mechanical Properties of a Weld Metal

The tensile machine used for these investigations was used at its maximum fixed speed, namely, 0·8 in. per minute. This speed is almost constant during the test.

The rate of straining depends primarily on the gauge length of the specimen, and will be only approximately constant during the test because of the extension of the specimen. For specimens of the type shown in Fig. 5(a), the rate of straining will vary from 0·025 per second to 0·015 per second during the test. The rate of straining will be correspondingly higher in specimens of the type shown in Fig. 5(c).

It is well known[4] that rate of straining is of considerable importance in determining tensile properties at elevated temperatures. Tests were therefore made at widely varying rates of strain on electrode A weld metal for testing temperatures of 1000 deg. C. and 1300 deg. C. The rates of strain used were approximately one-tenth normal and ten times normal. Results are reported in Table XIV. Both tensile strength and ductility become greater with increasing speed of testing. The two essential features, namely, low ductility at 1000 deg. C. and relatively high ductility at 1300 deg. C. are unaffected.

(m) Tensile Properties of Cast Steel

The behaviour of weld metal at elevated temperatures seemed to be unlike that of wrought mild steel, possibly as a result of melting during the welding process. If this is true cast steel should behave like weld metal.

Specimens were cut from two bars of cast mild steel (0·15 per cent. carbon), one cast at 1525 deg. C. and one at 1575 deg. C. These castings were made available by Mr. H. T. Protheroe, of Sheffield University, and had been tested for hot-tearing in a programme being carried out for B.I.S.R.A. His test consists essentially in casting steel bars in such a way that they are restrained in a jig on cooling, observations being made for cracking at high temperatures. For the steel under consideration, cracking had occurred at 1390 deg. C. after casting at 1575 deg. C., while no cracking had occurred in the steel cast at a lower temperature. From tests of this type, Protheroe[10] concludes that hot-tearing occurs at 1350 deg. C.-1450 deg. C. It is clear that hot-cracking in welds and hot-tearing in castings may be related phenomena, and from this point of view the tensile test would be expected to reveal corresponding brittleness at temperatures around 1400 deg. C.

The results of the tensile tests are reported in Table XV and the ductility values are plotted in Fig. 16.

For the steel cast at 1525 deg. C. the ductility curve shows the same general form as that for electrode A weld metal, which is shown in Fig. 16 for comparison—namely, a minimum in ductility at about 906 deg. C., with very high ductility at temperatures above 1200 deg. C. Values obtained for the steel cast at 1575 deg. C. are more erratic because of unsoundness, but even so they show the same general trend.

It is thus confirmed that with respect to tensile properties the behaviour of mild steel weld metal is like that of cast mild steel and unlike that of wrought mild steel. However, the surprisingly high ductility in the tensile test of reheated specimens of both cast steel and weld metal at temperatures above 1200 deg. C. is unexpected, and the point is given some attention in later discussion.

Table XI. Tensile Properties of Electrode A Weld Meta Deposited on "Padded" Plates

(Transverse specimens (Fig. 4b); dimensions* as shown in Fig. 5c)

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Reduction in Area (per cent.)
102	819	6·3	77
103	926	6·0	55
104	961	7·0	35
105	1004	6·6	36
106	1053	5·4	49
107	1104	4·3	73
108	1105	3·9	67
109	1195	3·1	92

* The gauge diameter was 0·18 ins., slightly smaller than shown in Fig. 5c, necessary because of slightly higher strength in the deposit.

Table XII. Tensile Properties of Two Electrode Core Wires

(Dimensions of gauge length 0·1 ins. diameter×0·5 ins. length)

Electrode A Core Wire

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Elongation per cent.	Reduction in Area per cent.
110	829	4·5	69	90
111	901	5·6	69	85
112	949	5·0	69	88
113	1003	4·5	68	87
114	1113	3·1	68	97
115	1212	2·4	73	99

Electrode B Core Wire

116	835	3·9	82	81
117	898	3·6	82	76
118	952	—	63	78
119	1013	4·6	62	75
120	1110	3·2	94	99

(n) Tensile Properties of an Austenitic Steel Weld

A few tests have been made on an austenitic steel weld on the 18/8 type. The results are reported in Table XVI, and the ductility values plotted in Fig. 17. The corresponding curve for electrode A weld metal is included in Fig. 17 for comparison. A minimum in ductility at about 1000 deg. C. is also obtained for the austenitic steel weld which is again general confirmation that the effect is a general characteristic of cast steel.

The values for tensile strength are plotted in Fig. 15; the strength values are much higher than those for mild steel for temperatures up to 1300 deg. C., the difference being more marked at the lower temperatures.

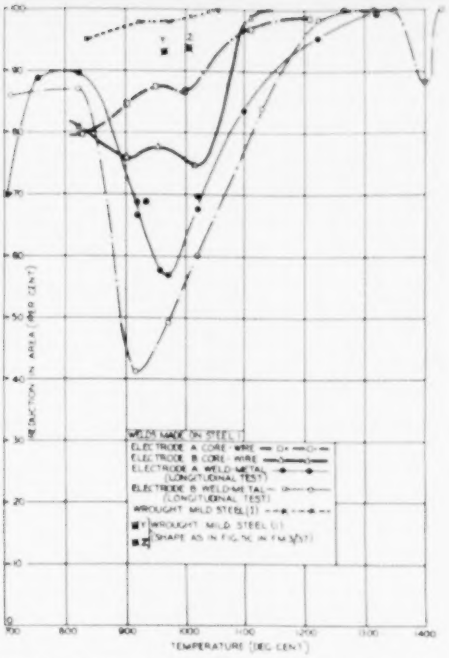
WELDS MADE ON STEEL 1
ELECTRODE A CORE WIRE
ELECTRODE B CORE WIRE
ELECTRODE A WELD METAL (LONGITUDINAL TEST)
ELECTRODE B WELD METAL (LONGITUDINAL TEST)
WROUGHT MILD STEEL (1)
WROUGHT MILD STEEL (1)
SHAPE AS IN FIG 9C IN FM 5-5
REDUCTION IN AREA (PER CENT)
TEMPERATURE (DEG CENT)

Fig. 14. Ductility of two weld metals and of their respective core wires.

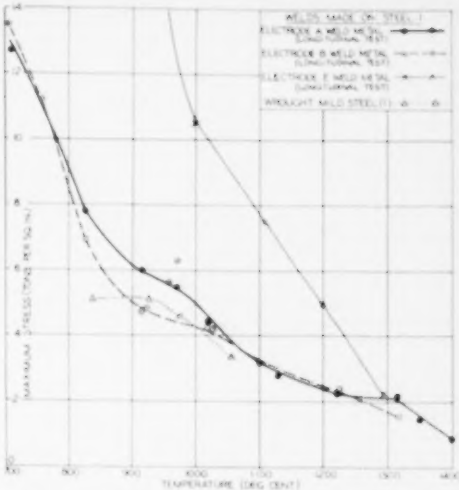
WELDS MADE ON STEEL 1
ELECTRODE A WELD METAL (LONGITUDINAL TEST)
ELECTRODE B WELD METAL (LONGITUDINAL TEST)
ELECTRODE E WELD METAL (LONGITUDINAL TEST)
WROUGHT MILD STEEL (1)
MAXIMUM STRESS (TONS PER SQ IN)
TEMPERATURE (DEG CENT)

Fig. 15. Tensile strength (maximum stress) of weld metal and wrought mild steel.

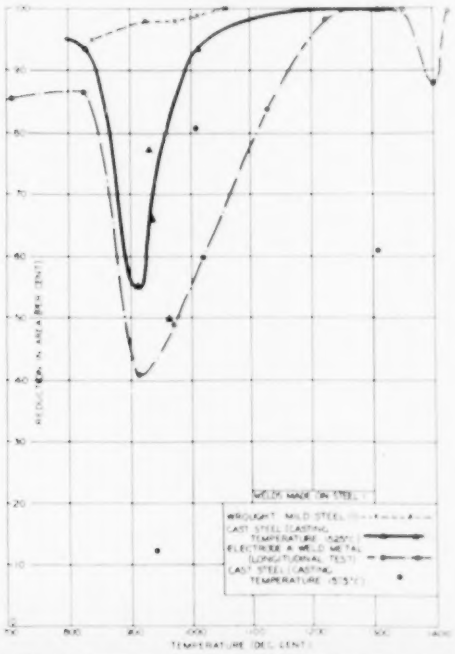
WELDS MADE ON STEEL
WROUGHT MILD STEEL (1)
CAST STEEL (CASTING TEMPERATURE 1525°C)
ELECTRODE A WELD METAL (LONGITUDINAL TEST)
CAST STEEL (CASTING TEMPERATURE 1575°C)
REDUCTION IN AREA (PER CENT)
TEMPERATURE (DEG CENT)

Fig. 16. Ductility of weld metal compared with that of wrought steel and cast steel.

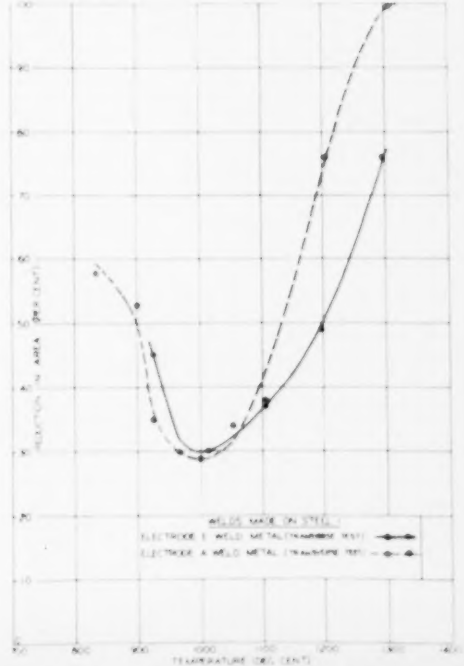
WELDS MADE ON STEEL
ELECTRODE E WELD METAL (LONGITUDINAL TEST)
ELECTRODE A WELD METAL (TRANSVERSE TEST)
REDUCTION IN AREA (PER CENT)
TEMPERATURE (DEG CENT)

Fig. 17. Ductility of austenitic steel weld.

Table XIII. Tensile Properties of Wrought Mild Steel (Steel 1)

(Dimensions as shown in Fig. 5b)

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Elongation per cent.	Reduction in Area per cent.
121	838	5·1	51	95
122	923	5·1	57	98
123	973	4·6	47	98
124	1010	3·9	88	99
125	1057	3·4	59	100

(Dimensions as shown in Fig. 5c)

126	963	6·2	—	93
127	1004	5·2	—	94

Table XIV. Effect of Rate of Straining on the Tensile Properties of Weld Metal (Electrode A/Steel 1)

(Transverse specimens (Fig. 4b), dimensions as in Fig. 5c)

Test No.	Testing Time (Seconds)	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Reduction in Area (per cent.)
138	2·5*	1001	8·8	64
29	25 (normal)	1000	5·1	29
139	225†	1007	3·9	28
140	3*	1302	4·8	100
34	28 (normal)	1307	1·8	99·5
141	270†	1318	1·4	90

* Tests made by rapid rotation of the hand-wheel of the tensile machine.

†Tests made by switching on and off at the lowest speed of the motor.

(o) Ductility of a High Carbon Steel

The tests on electrode A weld metal have shown very high ductility values at testing temperatures between 1200 deg. C. and 1425 deg. C. The latter figure represents the limit set by experimental difficulties, using the apparatus described here. It remains possible that low values of ductility would be encountered at higher temperatures. In particular low values might be expected above the solidus, which probably lies above 1490 deg. C.

With a high carbon steel it is possible to make tests above the solidus, since the latter occurs at a lower temperature. Tests have been made on a steel containing 0·85 per cent. carbon, which had a solidus temperature of about 1350 deg. C. Results are reported in Table XVII, and the ductility values have been plotted in Fig. 18. It is surprising that there is no sudden change in properties above the solidus, and that some strength and considerable ductility were recorded at temperatures as high as 1400 deg. C. In this connection some work of Hughes[3] has shown a similar effect in steel subjected to twisting at temperatures above the solidus.

Table XV. Tensile Properties of Cast Steel, Cast at Two Different Temperatures

(Dimensions as in Fig. 5b)

Steel cast at 1550 deg. C.

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Ton per sq.in.)	Elongation per cent.	Reduction in Area per cent.
128	832	5·0	70	94
129	914	—	36	55
130	937	4·1	28	66
131	947	3·7	54	77
132	966	5·1	36	50
133	1010	4·1	60	94
134	1309	1·4	84	100

Steel cast at 1750 deg. C.

135	941	4·1	8	12
136	1010	4·1	61	81
137	1309	1·3	84	61

Table XVI. Tensile Properties of a Weld from an Austenitic Electrode (Electrode E/Steel 1)

(Transverse specimens (Fig. 4b); dimensions of gauge length 0·13 in. diameter* × 0·125 in. length)

Test No.	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Reduction in Area (per cent.)
142	923	20·8†	46
143	1013	10·5	30
144	1108	7·5	37
145	1198	5·0	49
146	1295	2·2	76

* Due to the high strength of the austenitic weld, a smaller diameter than shown in Fig. 5c is necessary to ensure that fracture occurs in the weld.

† A still smaller diameter (0·10 in.) was required.

Table XVII. Tensile Properties of a High Carbon Steel

Steel 3 (0·85 per cent. Carbon)

Test	Testing Temperature (deg. C.)	Maximum Tensile Strength (Tons per sq.in.)	Elongation per cent.	Reduction in Area per cent.
147	955	4·7	77	99
148	1302	1·7	76	100
149	1351	1·2	103	100
150	1388	1·0	77	88
151	1397	0·9	24	58

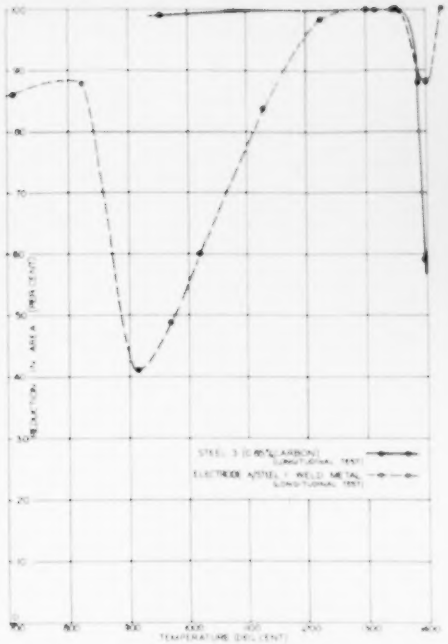

Fig. 18. Ductility of a high carbon steel (0·85 per cent. C.)

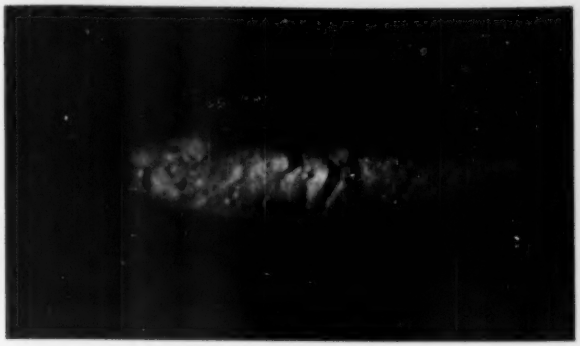

Fig. 19. Exploded cracks on the surface of a tensile specimen.

6. DISCUSSION

It has been found that mild steel weld metal exhibits a minimum in ductility somewhere in the temperature range 900-1100 deg. C. For two reasons this is probably due to some structural weakness in the weld metal. Firstly, the low ductility is more pronounced on testing transverse to the welding direction, i.e. across the plane of weakness produced by major segregation. Secondly, annealing presumably by causing homogenisation has a beneficial effect on the ductility. There is no direct evidence to suggest which particular factor is responsible for the minimum in ductility. Although the effect of iron oxide and sulphur has been investigated, the results, in general, are inconclusive, and while pick-up of impurities from the plate metal has been shown to reduce ductility the effect is not very marked.

Low ductility in the range 900-1100 deg. C. seems to be a general property of cast steel. The effect has been observed in an austenitic steel weld metal and also in a cast steel containing 0·2 per cent. carbon. Dupuy[6] found low values of ductility in the range 800-1100 deg. C. for two cast steels, containing 0·15 per cent. carbon and 0·44 per cent. carbon, and Piwowarski[7] obtained similar results for two cast steels containing 0·18 per cent. carbon and 0·44 per cent. carbon. Hodge[5] has reported a minimum in ductility at about 900 deg. C. for mild steel weld metal, though without commenting on the phenomenon. On the other hand, the effect is not observed in wrought steel. It has been shown that high ductility values are obtained in wrought mild steel over the temperature range 840-1050 deg. C. Ductility values are also much higher for the core wires than for their corresponding weld metals.

The range of low ductility for temperatures 900-1100 deg. C. is probably not connected with hot-cracking. The electrodes tested may be placed in the other D, C, B, A, for increasing susceptibility to hot-cracking, and in the order B, A, D, C, for increasing susceptibility to low ductility in the range 900-1100 deg. C. In addition, the use of an excessive welding current, which is known to promote hot-cracking, has a beneficial effect on ductility in this temperature range.

On raising the testing temperature from 1100-1200 deg. C. the ductility improves rapidly to give values approaching 100 per cent. reduction in area. Similar results were obtained by Dupuy[6] and Piwowarski[7] on cast steel. At first this was thought to be due to the annealing effect of the testing temperature, i.e. diffusion and homogenisation were removing the harmful structural defects causing brittleness at lower temperatures. This explanation was only partly confirmed by experiment. It was found that a prior treatment at 1340 deg. C. increased the ductility on subsequent testing at 1000 deg. C., although a short prior treatment at 1200 deg. C. had an adverse effect on ductility when tested at 925 deg. C.

The results for testing temperatures exceeding 1250 deg. C. are the most difficult to correlate with those of previously published work. Mild steel weld metal tested in the temperature range 1250-1425 deg. C. has exceedingly high ductility. Piwowarski[7] obtained quite different results. For several different steels he obtained a curve of characteristic shape with reduction in area descending from very high values at about 1250 deg. C. to very low values at about 1400 deg. C. The reason for this lack of agreement is not known.

Values for ductility are in general rather lower in Hall's[5] results than for those of other workers. This may be due to the fact that he used a different method of investigation, with tensile testing being carried out on cast specimens as they are cooling down from the molten condition. It suggested that the tensile properties of cast steel on reheating at very high temperatures may be different from the tensile properties of the steel on its initial cooling from the molten condition. This may explain why, on testing cast steel known to be susceptible to hot-tearing at about 1390 deg. C., very high values are obtained on reheating to testing temperatures exceeding 1300 deg. C.

Weld metal has been tested at temperatures up to 1425 deg. C., which represents the limit set by experimental difficulties with the apparatus described here. The solidus temperature lies well above this for mild steel (1490 deg. C.), but it has been found possible to test above the solidus for a high carbon steel (1350 deg. C.). In this case some strength and considerable ductility were found at temperatures 50 deg. in excess of the solidus temperature.

APPENDIX 1

Subsidiary Failure in Weld Metal Tensile Specimens

In one or two instances an unusual effect was observed on fractured all-weld-metal specimens. Near the main fracture a small crack would appear, on both sides of which the metal was distorted outwards (Fig. 19). The effect was only observed for testing temperatures around 1100 deg. C.

It may be that gas under pressure is responsible for the effect, though a second possible explanation cannot be dismissed, i.e. that uneven deformation in some cases sets up high tensile stresses in the surface layers.

APPENDIX 2

DETERMINATION OF THE PERCENTAGE AND CONSTITUTION OF THE OXIDE INCLUSIONS IN THREE SAMPLES OF MILD STEEL WELD METAL

The analyses were carried out in connection with the investigation into the tensile properties of mild steel weld metal at high temperatures being undertaken by the staff of the British Welding Research Association.

Samples of weld metal deposits from three different electrodes were forwarded to the N.P.L. and examined by the alcoholic iodine method, the residues, after solution of the metal, being analysed by micro-chemical methods and by X-ray diffraction. The three electrodes were

A. to BEAMA/IOW Classification E.217
C. to BEAMA/IOW Classification R.416
D. to BEAMA/IOW Classification E.537

The results obtained are recorded in Tables I and II. Table I shows the amounts of the various oxides and the total oxygen, expressed as a percentage of the metal, while Table II shows a comparison between the principal oxide constituents as determined by micro-chemical analysis and the compounds identified by X-ray examination of residues from each sample. It will be noted that two residues were prepared from each weld metal sample, one obtained by solution at 65 deg. C. for chemical analysis, the other by solution at 30 deg. C. for X-ray examination. The reason for the use of the lower temperature of solution in the second case is to ensure as far as possible that the oxygen-bearing compounds shall not be broken down, even partially, into their constituent primary oxides. At the same time the use of the lower temperature tends to retain in the residue some, if not all, of the sulphide inclusions, and thus enables a qualitative identification of their nature to be made.

It will be seen from Table II that in each sample there was a close correlation between the percentages of the oxide constituents determined chemically and the compounds identified by X-ray examination. The oxide inclusions in the deposit from electrode A contain a high proportion of SiO_2. Such inclusions would be expected to be rather glassy and to give a very weak X-ray diffraction pattern, which it was not possible to identify.* The presence in the sample of sulphur as MnS was, however, firmly established.

* The inability to identify the inclusions may have been because they contained significant amounts of TiO_2 and might therefore be expected to consist of a complex titanosilicate, for which no X-ray reference data are available.

Table I. to Appendix 2. Analyses of Alcoholic Iodine Residues from Welds at 65 deg. C.

Electrode BEAMA/IOW Classification	Code	Oxide in the Residue and Equivalent Oxygen Value Expressed as a Percentage of the Weld														Total Oxygen Per Cent Calculated
		SiO_2		FeO		Al_2O_3		MnO		Cr_2O_3		P_2O_5		TiO_2		
		Oxide	Oxygen	Oxide	Oxygen	Oxide	Oxygen	Oxide	Oxygen	Oxide	Oxygen	Oxide	Oxygen	Oxide	Oxygen	
E.217	A	0·088	0·047	0·013	0·003	0·005	0·002	0·084	0·019	0·005	0·002	0·004	0·002	0·018	0·007	0·082
R.416	C	0·082	0·044	0·009	0·002	0·002	0·001	0·164	0·037	0·001	0·0003	0·017	0·009	0·004	0·002	0·095
E.537	D	0·003	0·001	0·700	0·155	0·003	0·001	0·010	0·002	0·003	0·001	0·005	0·003	—	—	0·163

Table II to Appendix 2. Comparison between Analytical and X-ray Examination of Residues

Electrode BEAMA/IOW Classification	Principal Oxide Constituents of the Residue at 65 deg. C.			Compounds Identified by X-ray Examination of the Residue obtained at 30 deg. C.
	FeO per cent.	MnO per cent.	SiO_2 per cent.	
E.217	6	39	41 TiO_2 8	Small quantity of an unidentified compound + MnS
R.416	3 P_2O_5 6	59	29	2(Mn, Fe)O . SiO_2 +
E.537	97	1·0	0·4	Fe_3O_4 + small quantity Fe_2O_3 + FeS (probable)

Fig. 20. (×4700)

Fig. 21. (×4300)

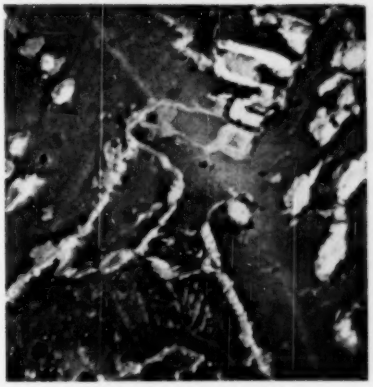

Fig. 22. (×4300)

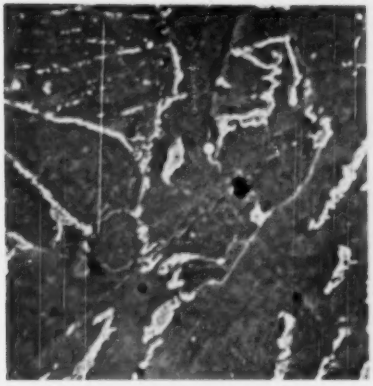

Fig. 23. (×4000)

Figs. 20-23. Structures found in electrode A weld metal.

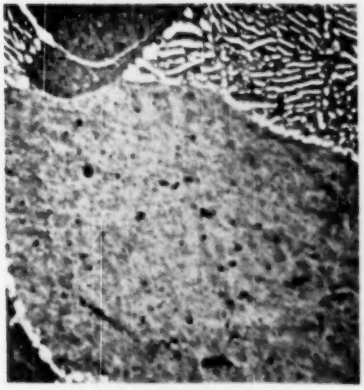

Fig. 24. (×4000)

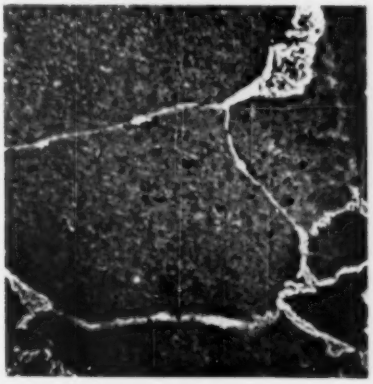

Fig. 25. (×4000)

Figs. 24-25. Structures found in electrode B weld metal.

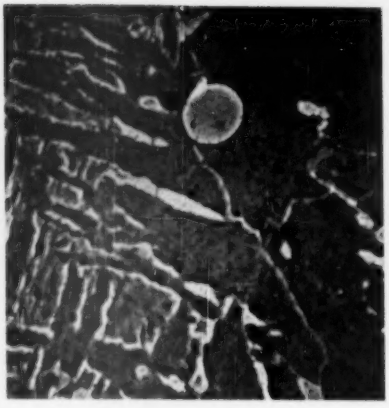

Fig. 26. (×4400)

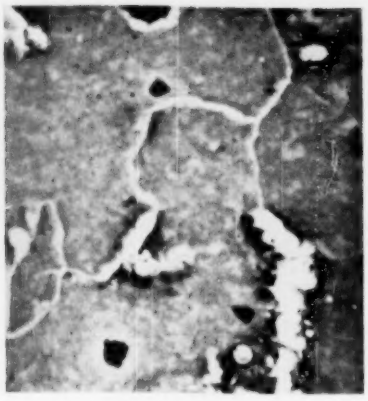

Fig. 27. (×5300)

Figs. 26-27. Structures found in electrode B weld metal.

Fig. 28. (×4400)

Fig. 29. (×4400)

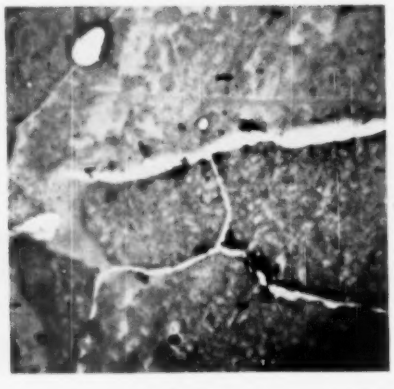

Fig. 30. (×4300)

Fig. 31. (×4400)

Figs. 28-31. Structures found in electrode D weld metal.

The sample from electrode C gave a much stronger X-ray diffraction pattern, the spacings in which were between those for tephroite ($2MnO . SiO_2$) and fayalite ($2FeO . SiO_2$) indicating that the inclusions consisted of tephroite in which a small portion of the manganese had been replaced by iron. In Table II this is indicated by the use of the formula $2(Mn, Fe)O . SiO_2$. No indication of the presence of a sulphide was found in the X-ray photograph of this residue.

Electrode D gives a deposit in which the oxygen is present essentially as iron oxide. This was confirmed by the X-ray examination. The presence of a small amount of Fe_3O_4 associated with the FeO may be noted. Other lines on the X-ray photograph indicated that the sulphur in this deposit is probably present combined with iron in the form of FeS.

APPENDIX 3

EXAMINATION OF WELD METAL STRUCTURES UNDER THE ELECTRON MICROSCOPE

In the course of investigations into the problem of hot-cracking in mild steel weld metal, some attention was given to the examination of the weld metal at high magnification under the light microscope. It was found that in the throat of the weld, i.e. in the region where hot-cracking is liable to occur, the grain boundaries appeared to be outlined with some constituent. This effect became more definite in weld metal liable to hot-cracking, but was in all cases on a very fine scale and could not be made out very clearly even at the limit of resolution of the light microscope, at a magnification of, say, 1,500 times.

It was felt that the effect might be better studied at even higher magnifications using the electron microscope. An investigation of this sort was kindly undertaken at the B.N.F.M.R.A. laboratories.

Accordingly, sections through three weld metals were submitted to them for examination:

Electrode A—A general purpose electrode, particularly susceptible to hot-cracking.

Electrode B—A general purpose electrode, not particularly susceptible to hot-cracking.

Electrode D—An oxide coated electrode, not at all liable to the defect of hot-cracking.

These electrodes are as used in the investigation on the mechanical properties at elevated temperatures, and the analyses of the weld deposits, together with the B.E.A.M.A. classification numbers for the electrodes, may be found in the section on "Materials" in the main body of the paper.

For each weld metal a replica was taken in the region of the throat of the weld, where hot cracking is most liable to occur. Examination under the electron microscope showed that, even in this small area, the structure was subject to wide variations. It was difficult, therefore, to cover these variations with a few typical photographs. In all, 52 photographs were taken from the replicas of the three weld specimens. Only 12 of these have been used in the present instance to illustrate particular points. Some of the written comment here must be taken as deriving from the experience gained by a general visual survey of the replicas under the microscope.

The technique used at the B.N.F.M.R.A. in obtaining the photographs was as follows:

The weld specimens were prepared in the usual way, except that the final polish was given using diamond powder. This technique has been described by E. C. W. Perryman.* They were then etched for 10-20 seconds in 3 per cent. nital solution. After cleaning the etched specimens by means of a collodion film, which was discarded, a thin Formvar replica film was applied. This was strengthened by a backing film of collodion and the double film was stripped under water. The backing film was subsequently dissolved, leaving the Formvar replica for examination.

Selected photographs, four from each replica, are given in Figs. 20-31, and illustrate the more interesting features observed. The magnifications used were in the range 4000-5000 times.

In the case of electrode A weld metal, which is specially liable to hot-cracking, the structures were in general rather coarse. For instance, Fig. 20 and, to a less extent, Fig. 21, show that the structure of the binary carbide entectoid is rather coarse by comparison with that from the other weld metals. It is more significant, however, that the grain boundaries are usually heavily outlined as shown in Fig. 22, though occasionally this feature is a little less pronounced (see Fig. 23).

The structures found in electrode B weld metal are much finer, Figs. 24 and 25 being typical. Occasionally coarser structures are found as shown in Figs. 26 and 27.

With electrode D weld metal, known to be free from hot cracking, the grain boundaries are much fainter and not always continuously outlined with any constituent. Figs. 28, 29 and 30 are typical. The binary carbide eutectoid is very fine and is small in amount (see Fig. 31), as would be expected since the carbon content is lower in this case.

Electrode D weld metal was found to contain the greatest quantity of non-metallic inclusions, though this is not shown in the photographs. The effect is in agreement with the results of the analyses reported in Appendix 2, which show that this weld metal had the highest iron oxide and total oxygen content.

Presumably the main constituent found on the grain boundaries is iron carbide, though it may be contaminated by traces of oxides and sulphides. Of course, any constituent of low melting point on the grain boundaries is suspect in connection with the problem of hot-cracking. It is interesting to note that, with the three weld metals examined, the extent to which they are susceptible to hot-cracking seems to bear some relation to the amount of precipitated constituent on the grain boundaries.

* E. C. W. Perryman, *Journ. Inst. Metals*, March, 1950, p. 61.

ACKNOWLEDGMENTS

Acknowledgments are gratefully made to:—

The British Non-ferrous Metals Research Association, for taking photographs of the structures of three weld metals with the electron microscope.

The National Physical Laboratory, for making the constitutional analysis on weld metal residues, and for making available their report on this work for inclusion as Appendix 2 in the present paper.

Dr. Reeve, for helpful advice and criticism, for making analyses and for supplying steel plate.
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Mr. Protheroe, for making available some cast steel for tests.

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R.64

Ductility of Mild Steel Welds near the Solidus

By E. C. Rollason, Ph.D., M.Sc., F.I.M.*

SYNOPSIS

The results of some simple hot tensile tests on core wire and weld metal show a lowering of ductility at 1250-1280 deg. C. Welds graded according to this criteria are in correct order as regards hot-cracking during welding. Increase of manganese in one type of electrode is shown to maintain hot ductility to a higher temperature.

INTRODUCTION

Hot-tearing of castings has been well known for a long time, and in welds a similar feature, known as hot-cracking, can occur. Whilst it is possible to design castings to minimise restrictions which would accentuate hot-tearing, it is almost impossible to avoid restrained joints in welded fabrication. Thus it is essential that electrodes should be capable of making welds free from hot cracking. Work with the Murex hot-crack machine, in which a single bead is strained during deposition by external force, had indicated that cracking occurred near the freezing temperature of the weld metal. Consequently, it was thought desirable to measure the change in ductility of mild steel weld metal at temperatures exceeding 1100 deg. C. This work was carried out in 1944 and relatively simple apparatus was used. The results, however, indicated a drop in ductility in the region of 1300 deg. C., and it was suggested to the R.F.M. Committee of B.W.R.A. in December, 1945, that in studying the causes of hot-cracking it would be worth while to determine the mechanical properties of weld metal near the solidus. As a result, Winterton and Ball have carried out some excellent work using refined apparatus, but the results failed to indicate any drop in ductility at 1200-1400 deg. C. It was thought desirable, therefore, to give a brief account of the early tests which draw attention to rate of heating as one factor which may be considered in future work.

APPARATUS

The simple form of hot tensile machine is shown in Fig. 1. This consisted of an upper fixed grip to take specimen 4 in. long with a gauge diameter of 0·191 in. The lower movable grip was attached to a simple lever system by means of which a load of 80 lb. could be applied suddenly to the specimen by rotating an eccentric cam. Very rapid failure ensued minimising the effect of creep and reduction of area measurements were used to indicate ductility, i.e.

$$\frac{(d_1 - d_2)^2}{d_1} 100 \text{ per cent.}$$

Resistance heating was used, the specimen being in the secondary of 2-volt transformer delivering up to 200 amps. The rate of heating to 1400 deg. C. was therefore a matter of seconds. To minimise oxidation a chamber, fitted with a

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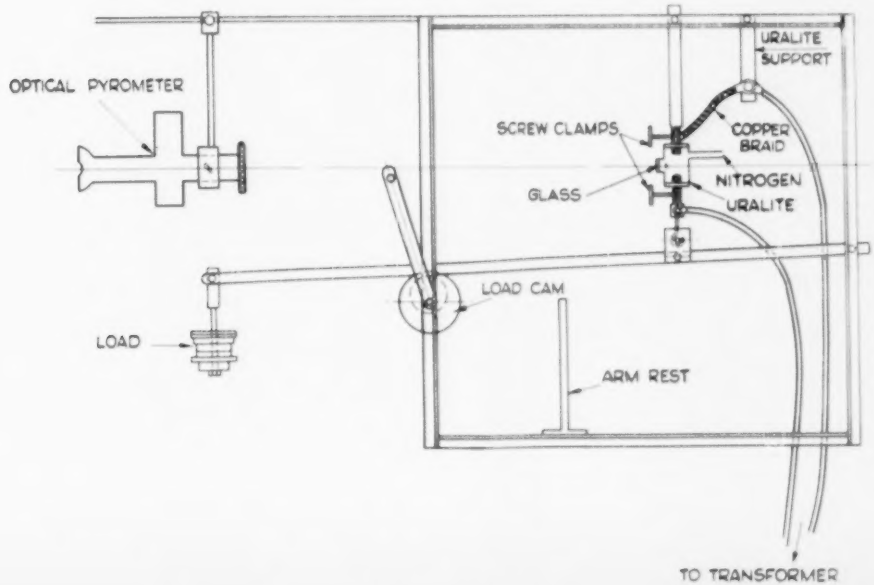

Fig. 1. Hot tensile tester.

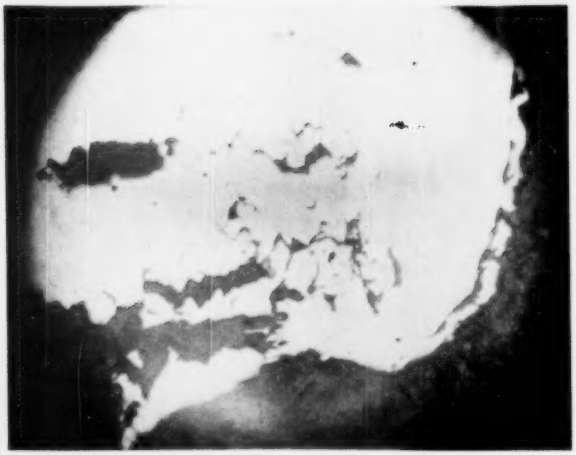

Fig. 3 *(right)*. General collapse in FR wire at 1290 deg. C.

×20

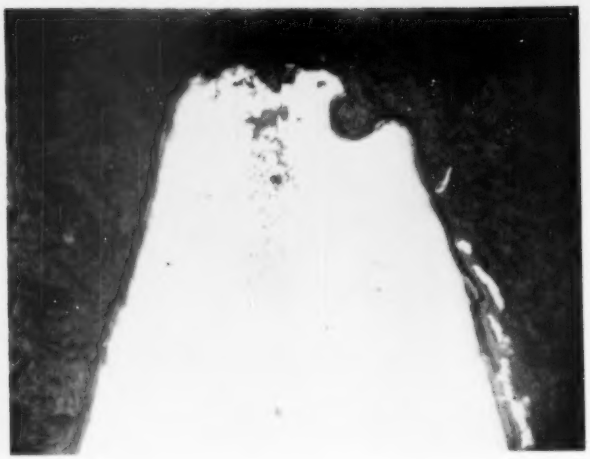

Fig. 4 *(left)*. Fracture of RR wire at 1290 deg. C.

30

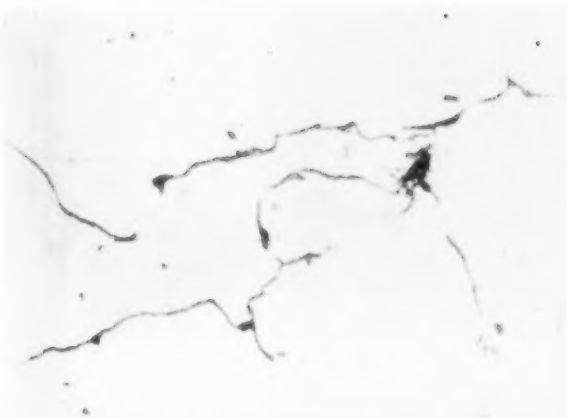

Fig. 5 *(right)*. Grain boundary collapse in FR wire at 1290 deg. C.

×70

glass observation window, surrounded the specimen and oxygen free nitrogen was used as a protective atmosphere, escaping around the lower grip. Temperature measurements were made by means of an optical pyrometer sighted on the wire specimen via the glass window. The readings were checked with a Pt-Pt/Rd couple on the specimen with reasonable agreement of ±10 deg. C.

RESULTS

Core Wire

Tests have been made both on core wires and on weld metal deposited from these core wires after coating with a single batch of flux. Typical results are given in Fig. 2 and analyses are given in Table I.

Table I. Composition of Core Wires and Weld Metal (R227 Type)

	Percentage					Hot Cracking
	C	Mn	Si	S	P	
F.R. Wire	0·13	0·39	0·01	0·034	0·049	
F.W. Weld	0·12	0·52	0·15	0·022	0·072	Bad
R.R. Wire	0·11	0·40	0·015	0·041	0·024	
R.W. Weld	0·10	0·53	0·16	0·028	0·042	Good

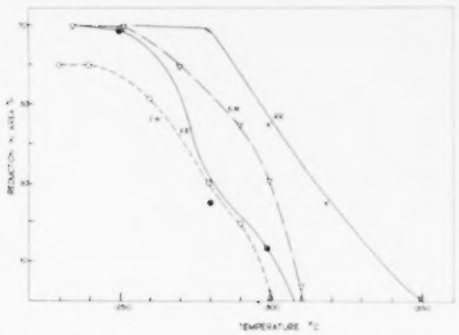

Fig. 2. Hot ductilities of two wires—FR, RR—and the welds from them with R227 coating.

From Fig. 2 it will be seen that the two wires begin to lose ductility at different temperatures, and this difference is reflected in the ductility and hot-cracking (on Murex Machine) of the weld metals. The ductility of the weld metal is somewhat lower than that of the wire. Since the coating is similar in each case, the variation in hot-cracking properties must be sought in variations in the wires.

Mode of Failure

The test specimen in the hot tensile test failed as a result of intercrystalline collapse of the austenite. If the ductility of the wire or weld is low there is very little necking and the specimen breaks off quickly. On the other hand, if the metal is ductile at its elevated temperature, necking occurs and the temperature rises before final collapse. The true temperature is, therefore, higher than that obtained in the test.

The temperature at which ductility begins to drop is likely to be reasonably accurate, but the temperatures for low values of reduction of area are higher than those reported.

Fig. 3 shows the general collapse of a wire having poor ductility (FR), while Fig. 4 shows the necking and internal failure of a typical ductile wire. Fig. 5 shows grain boundary cracks in a poor wire. The intergranular cracks in weld FW fractured at 1300 deg. C. are shown in Fig. 6.

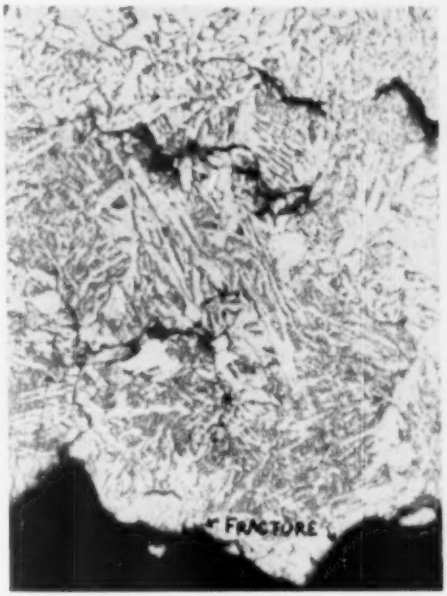

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Fig. 6. Intergranular cracks in FW weld metal at 1300 deg. C.

The Coating

The effect of poor wire on weld cracking varies with the type of coating and Table II indicates that some flux coatings, e.g. class 2, can minimise the effect to some extent.

Table II. Cracking Criteria of Various Coating Types Extruded on Wire of Low Hot Ductility (Murex m/c)

I.O.W.-B.E.A.M.A.	Coating	Current Amps.	Voltage	Hot Crack Index
Class 5	Iron oxide	215	24	40
Class 4	Iron-manganese silicate	195	25	26
Class 2	Titania	210	21	4

The results of hot tensile tests on all-weld metal samples made with bare wire and two fluxed electrodes are shown in Fig. 7. The E217 type weld metal shows superior ductility

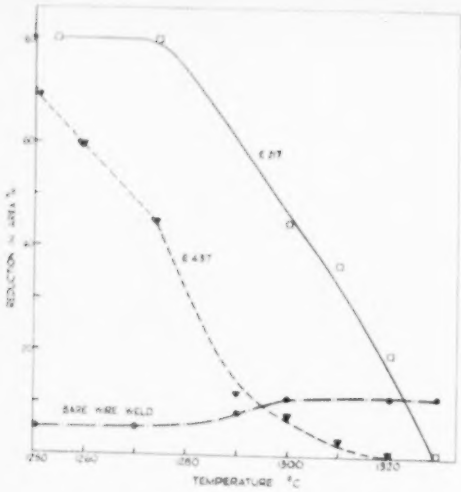

Fig. 7. Hot ductilities of three weld metals

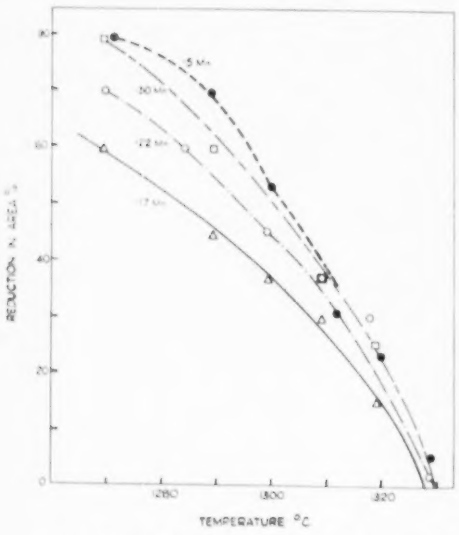

Fig. 8. Effect of manganese addition to weld via coating

over E437 type which agrees with practical welding experience in which this electrode cracks in a "T" fillet test. This is partly due to high carbon and low manganese contents. The bare wire weld has very low hot ductility and always cracks readily in the Murex hot-cracking machine.

The analyses of the welds in Fig. 3 are:

	C	Mn	S	P	N_2
Bare Wire ..	0·07	0·10	0·022	0·01	0·19
E437	0·16	0·24	0·026	0·05	0·02
E217	0·08	0·50	0·03	0·03	0·02

Although several factors influence the incidence of hot-cracking, it is recognised that elements which form low melting point films at the elevated temperature are pernicious. Sulphur as iron sulphide can readily induce hot-cracking of welds, but the addition of manganese can produce an improvement by forming high melting point MnS. To illustrate this point a series of electrodes were made using the same cast of wire and coatings with varying amounts of ferromanganese in the flux. The hot ductilities are shown in Fig. 8, and it will be observed that manganese additions to the coating of this electrode have beneficial effects as regards hot ductility of the welds.

DISCUSSION

The decrease in ductility of welds is found to start at about 1250 deg. C., although due to lack of refinement in the apparatus the true temperature may be somewhat higher. The lack of agreement between these results and those of Winterton and Ball may be due to either one or both of two differences in experimental procedure. The protective atmosphere would not be so effective as the vacuum used in the later work and the presence of some oxygen may have reduced the temperature of embrittlement. There is also a great difference in the rate of heating of the specimens. The few seconds heating used in the early tests would not allow much change to occur by diffusion, etc., at the elevated temperature, and this condition tends to approach that used in Hall's work in which tensile tests were carried out in specimens cooling from the molten condition, in which case intercrystalline films might have maximum effect. The slower heating used by Winterton and Ball allowed soaking to occur in the range 1000-1200 deg. C. and very fine films of sulphide might be expected to agglomerate and thus have less effect on ductility. Further, with low manganese contents the FeS which is likely to be formed near the solidus is likely, on soaking, to change to MnS, with less deleterious effect on hot-cracking.

This work was carried out in the Research Laboratories of Murex Welding Processes Limited and the practical help of P. Harris is acknowledged.

R.65

An Explanation of Hot-Cracking of Mild Steel Welds

By E. C. ROLLASON, Ph.D., M.Sc., F.I.M. and D. F. T. ROBERTS, B.Sc. (Tech.)*

SYNOPSIS

One explanation of the hot-cracking of mild steel welds is based on the major effect of sulphur and manganese, modified by the minor influences of phosphorus, silicon, carbon and nickel.

INTRODUCTION

Hot cracks in mild steel welds are intercrystalline tears occurring at an elevated temperature, and the surfaces of the crack are "blued" by oxidation. This appearance provides ready recognition from other forms of crack occurring near room temperature.

The cracks generally occur longitudinally in varying lengths in the bead, and sometimes as an extension of a crater crack (Fig. 1). Discontinuous branching cracks occasionally occur at right angles to the ripple marks (Fig. 2). A rarer form of crack occurs transversely (Fig. 3).

Cracking has been observed to develop while the arc is still playing on the crater, and a highly skilled welder can sometimes make a sound deposit from an electrode known to be somewhat susceptible to hot-cracking. This depends on the welder, mainly unconsciously, repairing minute cracks appearing in the semi-molten crater. This would be impossible unless cracks form near the melting point.

Hot cracks generally appear in a zone of relatively low strength which approximates to the throat region of the weld, the last part to solidify. As shrinkage of the cooler parts is occurring simultaneously the centre of the weld is subject to shrinkage stresses which produce plastic flow. In a sound weld the metal has sufficient ductility to flow as contraction takes place. A crack-susceptible weld has insufficient ductility and therefore tears.

Varying degrees of self-imposed strain occurs in practice. In a tee fillet joint some of the contraction of the bead on one side may be added to the self-contraction of the second bead; while in a ring weld around a tube in a plate the contraction of the first part of the bead is applied to the latter part which will be near the melting point of the metal. In multiple ring welds a complex stress system can be set up by the early welds which will cause a critical amount of strain in subsequent ones. Procedure of welding can therefore accentuate or minimise the risk of hot-cracking. Other factors are weld contour, welding current and arc length.

PREVIOUS WORK

Previous work has shown the causes of hot-tearing and hot-shortness of iron and steel to be carbon and sulphur, the latter only being important in hot-shortness since this applies to nominally completely solid material.

Norris[1] has shown that the relationship between manganese and sulphur, with a correction for the oxygen content, will explain the incidence of hot-shortness, due to FeS films. This work disposed of the theories based primarily on the effect of oxygen: certainly, if oxygen were the cause of hot-shortness, welds would be much more susceptible to it, and its elimination would be very difficult.

Hall[2] has shown both the effect of carbon and of sulphur as well as those of other materials, especially phosphorus, on the hot tensile properties of castings at temperatures close

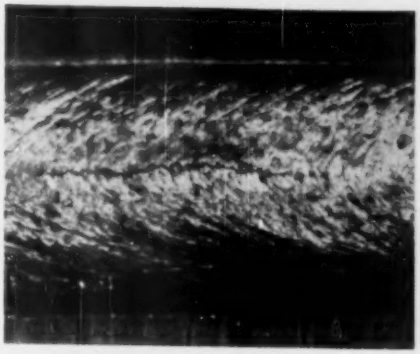

Fig. 1. Longitudinal hot crack.

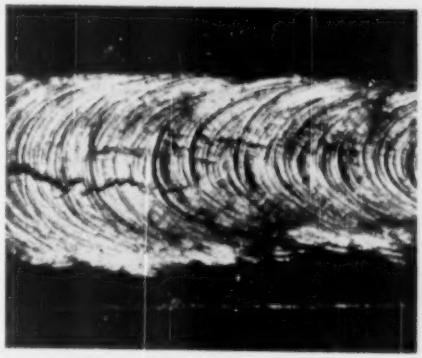

Fig. 2. Branching hot crack.

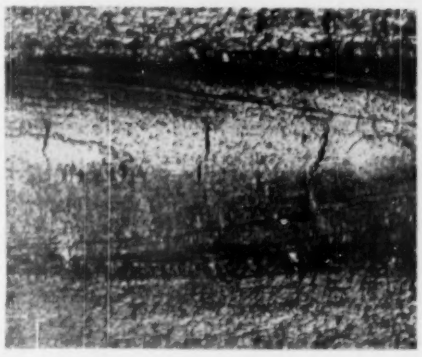

Fig. 3. Transverse hot crack.

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to the solidus. This work is, unfortunately, insufficiently detailed to give a complete account of the effects of S and P, but the effect of carbon is clearly shown and related to the constitutional diagram. It is clear that at over 1400 deg. C. only steels with less than 0·1 per cent. C., approximately, possess any useful ductility. It was also shown that increased S and P, in materials of similar analysis in other respects, caused a lowering of hot ductility.

Reeve[3] has shown that hot-cracking arising from the nature of the plate to be welded is associated with the sulphur pick-up from the plate, and has shown widely differing sensitivities to sulphur pick-up. This work, however, did not sufficiently distinguish between inherent hot-cracking of the weld metal and that due to sulphur pick-up, i.e. did not bring out clearly the amount of cracking obtainable with high quality plate due entirely to the electrode itself.

The importance of sulphur and the minor effect of phosphorus and carbon has been mentioned by Helin and Svantesson.[4]

A HYPOTHESIS

Crystallisation starts at the fused zone of the plate, and if the melt is not agitated by the electrode movement, long columnar crystals develop into the melt. In common with all alloys these crystals are purer than the deposited weld and the residual liquid is enriched in impurities. This concentration of lower melting point alloy occurs partly between the growing crystals and partly ahead of the crystal front, which finally forms an impure region near the throat of the weld.

Weld cracks may therefore be expected to occur mainly in the austenite boundaries (Fig. 4) and the plane between opposing crystal growths. Any element which forms low melting point compounds is, therefore, likely to increase hot shortness. Sulphur is such an element (Fig. 5 shows a film of FeS).

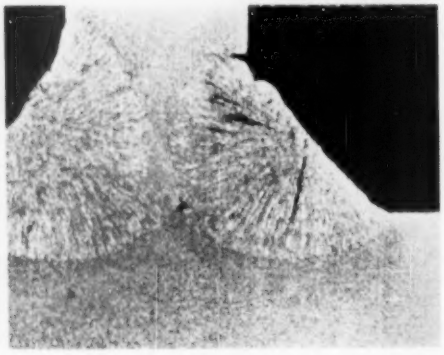

Fig. 4. Internal hot crack.

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Fig. 5. Sulphide film in weld metal.

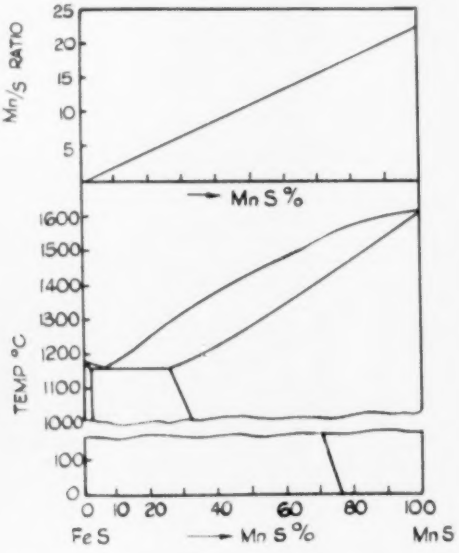

Fig. 6. Equilibrium diagram of the FeS-MnS systems and the relation of amount of MnS to the Mn/S ratio.

Sulphur

For many years sulphur has been recognised as an important factor in hot tearing or cracking. The reason is readily seen from Fig. 6 which gives the melting point of FeS as 1193 deg. C. The presence of traces of the FeS around the austenite boundaries reduces cohesion even at 1200 deg. C. It has been known for many years that the addition of sufficient manganese to a steel improves hot ductility by forming MnS or managnese rich sulphides with melting points above that of steel. MnS becomes increasingly stable as the temperature drops to 900 deg. C., and MnS can also dissolve a large amount of FeS as shown by the solubility curve in Fig. 6. Whiteley[7] has stated that in order to form MnS at the freezing point of steel the ratio of manganese to sulphur should exceed 22. The steel maker, however, often avoids hot tearing trouble even with a ratio of Mn/S = 5. In welds, on the other hand, it is desirable to maintain a ratio of 15. Much discussion in the past has been concerned with hot shortness in forgings and it is desirable to consider how the conditions in such an application vary from that found in weld metal.

The steel ingot cools slowly and is not subjected to much restraint and, before rolling, is homogenised to some extent in the soaking pit. As a result manganese has a reasonable chance of combining with sulphur and sulphide films tend to ball-up, in which form they have less effect on the cohesion.

In comparison a weld cools much more rapidly than an ingot and sulphur tends to be retained in solution. This is readily demonstrated by taking sulphur prints of a normal weld and one annealed at 1200 deg. C. The former weld gives a pale print indicating little massive MnS to be attacked by the acid. With rapid cooling therefore, any excess of sulphur over the solubility limit will form FeS due to insufficient time for diffusion processes to occur. Solubility of FeS appears to be approximately 0·3 per cent. at 1250 deg. C.

Reeve has demonstrated the effect of sulphur pick-up from the parent plate. The effect of sulphur from the flux coating is shown as follows:

Electrode	*Amount of Sulphur in Coating*	*Length of Hot Crack*
E.317	Trace	0
	0·04	1½ in.

Effect of sulphur in the wire on the cracking of various types of weld metal is shown in Table I, in which the length of the test specimen was 3 in. The Murex hot crack machine was used.

Table I

Composition of Wire				Length of Hot Crack, Inches			
S	Mn	C	Si	E.517	E.317	E.217	E.645
0·023	0·25	0·08	Tr.	2	1½	0	0
0·035	0·25	0·08	Tr.	2	1½	0	0
0·046	0·26	0·08	Tr.	2*	1½	1	½

* No cracking in normal service, see later explanation.

DEAD SOFT WELDS

Dead soft mild steel weld metal produced from coatings rich in iron oxide offer, under certain conditions, a tolerance for sulphur without cracking. Up to about 0·06 per cent. sulphur it is considered that the FeS formed is combined with the isolated particles of iron oxide rich inclusions and, therefore, has less effect on boundary weakening. Above about 0·06 per cent. sulphur FeS (or FeS-FeO eutectic) films form and cracking susceptibility increases. In the rutile E.317 type electrode the omission of manganese from the coating gives weld metal with a fairly high resistance to hot-cracking under certain conditions of cooling, these being "soft" weld metals, not "dead soft."

MANGANESE

The addition of manganese to weld metal reduces the amount of FeO inclusions (Sloman[6]), and tends to form (MnFe)S, but from Fig. 4 it will be seen that about 60 per cent. MnS is necessary to raise the melting point of the inclusion to 1500 deg. C. A mixed sulphide containing 25 per cent. MnS has such a low melting point that grain weakening films are still likely to cause hot-cracking.

The first effect of a manganese addition to oxidised weld metal is therefore likely to increase cracking while further additions will reduce cracking. This is in agreement with practical experience in electrode development. In considering the effect of manganese it must be borne in mind that manganese as estimated by the analyst may exist in different forms with varying effects, e.g. MnO, MnS, $2MnOSiO_2$, or Mn. As regards effect on hot cracking, Mn is of prime importance.

PREHEATING

Slowing the rate of cooling by preheating can give contradictory results as regards improvement in hot-cracking tendency depending on the constitution of the particular weld metal. With intermediate manganese content in association with iron-manganese silicate fluxes, preheating can increase hot-cracking by allowing manganese to remove sulphur from solution so increasing the amount of (FeMn)S of low melting point, which then forms films.

With E.217 type high rutile electrodes (with $2MnOSiO_2$ inclusions), preheating will again increase the amount of MnS, but the melting point of the sulphides is likely to be raised to such a point that tendency to hot-crack is not increased, and some reduction is likely due to the lower contraction.

HIGH CURRENT

It is well known that as the current value approaches the maximum tolerated by the electrode hot-cracking tendency increases. Although the retarded cooling obtained by the use of high current values will improve the resistance of E.217 welds, this advantage is more than balanced by the superheating of the weld which results in larger crystals and less grain boundary area to accommodate sulphide.

SILICON AND PHOSPHORUS

In agreement with Reeve, silicon and phosphorus are considered to adversely affect hot-cracking as shown by the following results using different wires and the same batch of flux.

Table II

Effect of	Code	Weld Metal Analyses					Hot Crack Specimen 3" long
		C	Mn	S	P	Si	
Phosphorus	S	0·13	0·55	0·02	0·039	0·15	0
	F	0·125	0·51	0·022	0·072	0·15	1·7 in.
Silicon	Y	0·11	0·50	0·026	0·056	p·135	0
	R	0·10	0·53	0·128	0·042	0·162	2 in.

The variation of silicon and phosphorus content to produce a given change in hot-cracking quality is greater than that of sulphur by a factor sufficiently large to suggest that these elements are not themselves primarily responsible for cracking. It is considered that they act in a secondary manner by increasing the segregation of sulphur as mentioned by Whiteley,[5] who states that at a given manganese content the segregation of carbon, phosphorus and sulphur increases with the total content of sulphur, phosphorus and carbon. Such increased segregation of sulphur is likely to increase the risk of the formation of weak films. The effect of silicon will depend on the manganese content of the weld, but is thought to affect segregation of sulphur in a similar manner

to that of phosphorus. Below 0·1 per cent. silicon little metallic silicon is likely to be present in the weld. With silicon increases to between 0·1 to 0·2 per cent. it is probable that the amount of metallic manganese in the weld, even though total amount is the same, will increase, with consequent improvement in hot cracking. Above these percentages more deleterious effects are usually found if Mn is not simultaneously increased. Phosphorus is more potent than silicon since little loss occurs in welding, whereas silicon can be oxidised under certain conditions.

CARBON

The carbon content can be an important factor in hot-cracking phenomena. Reference to the work of Hall[5] shows that the ductility of steel falls rapidly as the carbon content rises from 0 to 0·1 per cent. which covers the range of welds from the dead soft E.537 to E.217. This fall in ductility is associated with a rapid development of strength which will increase the self-imposed restraint on the weld metal as its carbon content rises.

For example, highly oxidised weld metals are not prone to cracking in practice, but can be readily cracked on the Murex machine where an independent amount of strain can be applied. In other words, in practical applications the dead soft weld metal does not develop enough strain to cause dangerous cracking.

NICKEL

Nickel in amounts as low as 0·5 per cent. in a weld can materially increase hot-cracking, e.g.:

Cracking Length (Ni nil)	*Wire Ni Content* %		*Cracking Length, with Ni*
$\frac{1}{4}$ in.	E.317	0·08	$\frac{1}{2}$ in.
$\frac{1}{2}$ „		0·32	$1\frac{3}{4}$ „
1 „		0·60	3 „

	Coating Addition Weld Ni per cent.	
Low Mn (0·35) deposit 1 in.	0·5	3 in.
High Mn (0·80) deposit 0 „	1·5	$1\frac{1}{2}$ „

Although the complete explanation of the effect of nickel is not understood, it is associated with presence of sulphide films around the grain boundaries of γ. In a single bead these films consist largely of FeS depending on the type of electrode. In, for example, high manganese (E.645 coating) welds a greater quantity of MnS is present. In multi-bead welds these sulphide films are dove grey in colour and respond in etching tests similarly to manganese sulphide. They resist balling up on heat treatment at 1000 deg. C. and in the main appear to be similar to the sulphide films found in pure nickel by Hall.[5] When quenched from 1200 deg. C. some product of indeterminate composition is observed, being dark brownish grey in colour and very powdery on polishing.

It would appear, therefore, that for a given sulphur content in a weld nickel is instrumental in increasing the amount of grain boundary sulphide and consequent cracking.

This work was carried out in the Research Laboratories of Murex Welding Processes Limited, Waltham Cross, Herts.

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